



# STEEL AND ITS HEAT TREATMENT

*Volume II—Engineering and Special-  
Purpose Steels*

BY  
D. K. BULLENS

THE METALLURGICAL STAFF  
*Of the Battelle Memorial Institute*

*FOURTH EDITION, REWRITTEN AND RESET*

NEW YORK  
JOHN WILEY & SONS, INC.  
LONDON: CHAPMAN & HALL, LIMITED



IN THE REPRINTING OF THIS BOOK, THE RECOMMEN-  
DATIONS OF THE WAR PRODUCTION BOARD HAVE  
BEEN OBSERVED FOR THE CONSERVATION OF PAPER  
AND OTHER IMPORTANT WAR MATERIALS. THE  
CONTENT REMAINS COMPLETE AND UNABRIDGED.

COPYRIGHT, 1916, 1917  
BY DENISON K. BULLENS

COPYRIGHT, 1927  
BY D. K. BULLENS  
COPYRIGHTED IN GREAT BRITAIN

COPYRIGHT, 1935  
BY DENISON K. BULLENS

COPYRIGHTED CANADA, 1935  
INTERNATIONAL COPYRIGHT, 1935

COPYRIGHTED CANADA, 1939  
INTERNATIONAL COPYRIGHT, 1938

COPYRIGHT, 1939  
BY BATTELLE MEMORIAL INSTITUTE

*All Rights Reserved*  
*This book or any part thereof must not*  
*be reproduced in any form without*  
*the written permission of the publisher*

PRINTED IN U. S. A.

## PREFACE

This is the second volume of the revision of "Steel and Its Heat Treatment." So much information, too important to omit, has accumulated since the last revision of the book that a single volume would be too bulky. The principles, processes, and fundamentals of control of procedures vital to heat treatment have been discussed in Volume I. That discussion refers to steel in general, alloy steel as well as plain carbon steel, without detailed consideration of the effect of alloying elements.

This volume deals with engineering and special purpose steels, both carbon and alloy.

It has been well said that this is the day of "tailor-made" steel, steel whose composition and heat treatment are adjusted to fit it for some specialized engineering use, not adequately served by tonnage steels.

In order that other special steels may be made available as new and different needs arise, and that intelligent choice may be made among the compositions of, and treatments for, those now available, the characteristics of the alloying elements, the special properties they confer upon steel, and the variations in heat treatment required to make the fullest and most economical utilization of the alloy steels have been given special attention. Cast steel has been given equal consideration with wrought steel.

The cooperation of the same individuals, technical societies, firms, and members of the Battelle staff, acknowledged in the preface to Volume I, has been equally helpful in the preparation of this volume. Special acknowledgment is made to the Society of Automotive Engineers for data included in this volume.

All temperatures mentioned herein are in the Fahrenheit scale, and the  $^{\circ}\text{F}$  is omitted;  $1000^{\circ}$  means  $1000^{\circ}\text{F}$ . A Fahrenheit-Centigrade conversion table is included for those who wish to make the conversion.

Mechanical properties are given in U. S. units, and to save space are abbreviated; e.g., instead of "tensile strength, 100,000 lb./in.<sup>2</sup>, yield strength 80,000 lb./in.<sup>2</sup>, elongation 25 per cent in 2 inches, reduction of area 60 per cent, Izod impact 55 ft.-lb.," the notation will be "tensile 100,000, yield 80,000, elong. 25%, R.A. 60%, Izod 55." Similarly the name of an alloying element which recurs frequently in a discussion will not be spelled out; the chemical symbol will be used.

"Yield strength" is used throughout instead of "elastic limit" in accordance with the terminology of the American Society for Testing Materials.

D. K. BULLENS

CLYDE E. WILLIAMS, Director  
Battelle Memorial Institute.

# TEMPERATURE CONVERSION TABLE

By DR. LEONARD WALDO

Reprint from Metallurgical and Chemical Engineering

C. °	0	10	20	30	40	50	60	70	80	90		
	F. °	F. °	F. °	F. °	F. °	F. °	F. °	F. °	F. °	F. °		
-200	-328	-346	-364	-382	-400	-418	-436	-454	-472	-490		
-100	-148	-166	-184	-202	-220	-238	-256	-274	-292	-310		
0	+ 32	+ 14	- 4	- 22	- 40	- 58	- 76	- 94	- 112	- 130		
0	32	50	68	86	104	122	140	158	176	194	C. °	F. °
100	212	230	248	266	284	302	320	338	356	374	1	1.8
200	392	410	428	446	464	482	500	518	536	554	2	3.6
300	572	590	608	626	644	662	680	698	716	734	3	5.4
400	752	770	788	806	824	842	860	878	896	914	4	7.2
500	932	950	968	986	1004	1022	1040	1058	1076	1094	5	9.0
600	1112	1130	1148	1166	1184	1202	1220	1238	1256	1274	6	10.8
700	1292	1310	1328	1346	1364	1382	1400	1418	1436	1454	7	12.6
800	1472	1490	1508	1526	1544	1562	1580	1598	1616	1634	8	14.4
900	1652	1670	1688	1706	1724	1742	1760	1778	1796	1814	9	16.2
1000	1832	1850	1868	1886	1904	1922	1940	1958	1976	1994	10	18.0
1100	2012	2030	2048	2066	2084	2102	2120	2138	2156	2174		
1200	2192	2210	2228	2246	2264	2282	2300	2318	2336	2354		
1300	2372	2390	2408	2426	2444	2462	2480	2498	2516	2534		
1400	2552	2570	2588	2606	2624	2642	2660	2678	2696	2714	F. °	C. °
1500	2732	2750	2768	2786	2804	2822	2840	2858	2876	2894	1	.56
1600	2912	2930	2948	2966	2984	3002	3020	3038	3056	3074	2	1.11
1700	3092	3110	3128	3146	3164	3182	3200	3218	3236	3254	3	1.67
1800	3272	3290	3308	3326	3344	3362	3380	3398	3416	3434	4	2.22
1900	3452	3470	3488	3506	3524	3542	3560	3578	3596	3614	5	2.78
2000	3632	3650	3668	3686	3704	3722	3740	3758	3776	3794	6	3.33
2100	3812	3830	3848	3866	3884	3902	3920	3938	3956	3974	7	3.89
2200	3992	4010	4028	4046	4064	4082	4100	4118	4136	4154	8	4.44
2300	4172	4190	4208	4226	4244	4262	4280	4298	4316	4334	9	5.00
2400	4352	4370	4388	4406	4424	4442	4460	4478	4496	4514	10	5.56
2500	4532	4550	4568	4586	4604	4622	4640	4658	4676	4694	11	6.11
2600	4712	4730	4748	4766	4784	4802	4820	4838	4856	4874	12	6.67
2700	4892	4910	4928	4946	4964	4982	5000	5018	5036	5054	13	7.22
2800	5072	5090	5108	5126	5144	5162	5180	5198	5216	5234	14	7.78
2900	5252	5270	5288	5306	5324	5342	5360	5378	5396	5414	15	8.33
3000	5432	5450	5468	5486	5504	5522	5540	5558	5576	5594	16	8.89
3100	5612	5630	5648	5666	5684	5702	5720	5738	5756	5774	17	9.44
3200	5792	5810	5828	5846	5864	5882	5900	5918	5936	5954	18	10.00
3300	5972	5990	6008	6026	6044	6062	6080	6098	6116	6134		
3400	6152	6170	6188	6206	6224	6242	6260	6278	6296	6314		
3500	6332	6350	6368	6386	6404	6422	6440	6458	6476	6494		
3600	6512	6530	6548	6566	6584	6602	6620	6638	6656	6674		
3700	6692	6710	6728	6746	6764	6782	6800	6818	6836	6854		
3800	6872	6890	6908	6926	6944	6962	6980	6998	7016	7034		
3900	7052	7070	7088	7106	7124	7142	7160	7178	7196	7214		
C. °	0	10	20	30	40	50	60	70	80	90		

EXAMPLES: 1347° C. 2444° F. + 12°.6 F. = 2456°.6 F.: 3367° F. = 1850° C. + 2°.78 C = 1852°.78 C.



# CONTENTS

## VOLUME II. ENGINEERING AND SPECIAL-PURPOSE STEELS

### *SECTION I. THE SIMPLE ENGINEERING STEELS*

CHAPTER	PAGE
1. CARBON STEELS . . . . .	1
2. PHOSPHORUS, SULPHUR, LEAD, AND FREE-MACHINING STEELS . . . . .	49
3. MANGANESE CONSTRUCTIONAL STEELS . . . . .	61
4. SILICON STEELS . . . . .	84
5. COPPER STEELS . . . . .	99
6. THE PLAIN NICKEL STEELS . . . . .	113
7. THE PLAIN CHROMIUM STEELS . . . . .	130
8. THE ASSISTANT ELEMENTS, ALUMINUM, TITANIUM, ZIRCONIUM, VANADIUM, ETC. . . . .	157
9. PLAIN MOLYBDENUM AND TUNGSTEN STEELS . . . . .	168
10. COMPARISON OF ALLOYING EFFECTS: S.A.E. STEELS AND TREATMENTS . . . . .	177

### *SECTION II. ENGINEERING STEELS OF MORE COMPLEX COMPOSITION*

11. HIGH-YIELD STRENGTH, LOW-ALLOY STEELS . . . . .	207
12. COMPLEX ALLOY STEELS BASED ON FERRITE FORMERS. LARGE SECTIONS. THE NICKEL-CHROMIUM AND SIMILAR FAMILIES. . . . .	221
13. COMPLEX ALLOY STEELS BASED ON CARBIDE FORMERS. CHROMIUM AND SIMILAR FAMILIES . . . . .	251
14. ALLOY STEELS FOR CARBURIZING . . . . .	271
15. ALLOY STEELS FOR NITRIDING . . . . .	294
16. SPRING STEELS . . . . .	302
17. STEEL CASTINGS . . . . .	316
18. FACTORS IN THE SELECTION OF CONSTRUCTIONAL STEELS . . . . .	321

### *SECTION III. THE SPECIAL-PURPOSE STEELS*

19. STEELS FOR HIGH- AND LOW-TEMPERATURE SERVICE . . . . .	336
20. THE AUSTENITIC STEELS . . . . .	396
21. CARBON TOOL STEELS . . . . .	420
22. LOW AND INTERMEDIATE ALLOY TOOL AND DIE STEELS . . . . .	433
23. HIGH-SPEED TOOL STEELS . . . . .	444
24. PERMANENT MAGNET STEELS AND THEIR HEAT TREATMENT . . . . .	460
INDEX . . . . .	471



# STEEL AND ITS HEAT TREATMENT

## *SECTION I. THE SIMPLE ENGINEERING STEELS*

### CHAPTER 1

#### CARBON STEELS

**Introduction.** The principles of heat treatment and a knowledge of the reasons for the operations involved in heat treatment, discussed in Vol. I, can be applied to the handling of a wide variety of C, of simple, and of complex alloy steels to produce steel parts ready for engineering use. The engineer must choose some steel and some one treatment from among the various possibilities. The engineering phase requires an understanding of the behavior of simple and complex steels under varying heat treatments, and of the possibilities and limitations in the range of properties thus securable. Replacement of one steel by another of different composition and treatment is, ordinarily, entirely feasible. Sometimes, however, so peculiar a property is required that the alternative steels and treatments are few indeed.

The steels that enter into engineering construction, as parts of machines, engines, and the like, are generally chosen for some optimum combination of machinability, cold formability, and strength and toughness at atmospheric temperatures. Some of these properties are not readily determined or expressed, but in general they can be approximated from the results of the common tensile, hardness, impact, and endurance tests. As a rule, any such part could be made from a wide variety of steels, with suitable heat treatment chosen for each. Within this range of possibilities, economic conditions, i.e., the cost of the steel itself, and of the requisite heat treatment and machining, are the deciding factors. The steels used in construction may, therefore, be more conveniently considered under a broad grouping within which they may be compared, very largely on the basis of standard tests, leaving the special-purpose steels for consideration in a later section.

**Size of Section.** In selecting a steel and a heat treatment, if heat treatment is required, to make a particular part for a definite service.



the engineer must balance many factors. The properties specified can often be obtained by choosing any one of a wide variety of C, of simple alloy, or complex alloy steels, if the heat treatment is suitably varied, *provided the part to be made is small*. As the section increases, and C steels become inadequate, the variety of alloy steels amenable to practical heat treatment decreases, but a choice is still available.

**Experience and Judgment.** All too often the properties specified in terms of the usual tests do not adequately evaluate the factors that really enter into performance so that the common tests may lead to the conclusion that two steels are equal when they are not. In such cases experience needs to be drawn upon—one's own, that of other users and that of the steel producer, who is as anxious as is the user that the proper choice be made. However, a preliminary classification, into obviously unsuitable and possibly suitable steels, can be made on the basis of standard tests and known propensities.

Such factors as uniformity, cost, availability as standard products without having to have a special heat made, differences in machinability, and amenability to other fabricating processes without too radical change in plant practice, distortion in hardening, differences in notch susceptibility under impact or repeated stress, etc., will weigh heavily in still further narrowing the choice among those steels that might serve. Individual conditions, which shift from time to time, vary so that there will always be room for sound engineering judgment based on much that cannot be set down in texts, charts, and tables.

In the discussion that follows the outstanding properties of the most-used steels will be presented and some attention paid to other steels that might be of value but are not yet standard products. The range of possible alloy combinations and of heat treatment is so great that space is lacking to set down anything like a complete summary of the information on record for each. Nor is this necessary since there are modern books and monographs which collect the previously scattered information for most of the major alloying elements, and handbooks in which data are concisely summarized. The effort will be to present a unified picture rather than a detailed one.

**Classifications.** The possibilities and limitations of the various alloys in steels for heat treatment might be discussed with reference to the effects upon the

ferrite

carbide

critical range on heating

properties as normalized—especially in cast steels

**S.A.E. Carbon Steels.** The Society of Automotive Engineers has instituted a numbering system for steels which serves as a very convenient method of designation of the commonly used C and alloy steels.

The C steels included in the S.A.E. list are:

	C%	Mn%		C	Mn
1010	0.05-0.15	0.30-0.60	1075	0.70-0.85	0.60-0.90
1015	0.10-0.20	"	1080	0.75-0.90	"
1020	0.15-0.25	"	1090	0.85-1.00	"
1025	0.20-0.30	"	1095	0.90-1.05	0.25-0.50
1030	0.25-0.35	0.60-0.90	X1040	0.35-0.45	0.40-0.70
1035	0.30-0.40	"	X1045	0.40-0.50	0.40-0.70
1040	0.35-0.45	"	X1050	0.45-0.55	0.40-0.70
1045	0.40-0.50	"	S Max.	0.04	
1050	0.45-0.55	"	P Max.	0.055	
1055	0.50-0.60	"	Si {	Min. 0.15 All	
1060	0.55-0.70	"		Max. 0.30 in basic O. H., not	
1065	0.60-0.75	"		limited in acid O. H. and	
1070	0.65-0.80	"		electric.	

The Mn range shifts from 0.30-0.60% to 0.60-0.90% at S.A.E. 1030. This Mn increase has something to do with the trend of the curves in Fig. 1. From 0.35 to 0.55% C the "X" series is also available in which the increase in Mn over that in the lower C steels is held down to 0.10%. Also designated by the X prefix are a few others with 0.70-1.00 and 0.90-1.20% Mn, which in commercial practice are classed as C steels, but from the metallurgical point of view are Mn steels, and are discussed herein under that heading. In S.A.E. 1095 steel, the Mn is cut to 0.25-0.50% because this grade is essentially a tool steel (see Chap. 21).

The effect of these variations in Mn will be brought out in Chap. 3. It will be noted that the Si and Mn contents are high enough so that these S.A.E. steels are all *killed* steels, whereas much of the low-C sheet steel used for cans, automobile bodies, and the like is *rimmed* steel. The S.A.E. C steels are supposed to be produced with great care as to quality, analogous to alloy steels.

**Size Effect in Wrought Steel.** Although the effect of section size is nowhere near so marked in normalizing, and annealing as it is in quenching and tempering, it does count. Primarily owing to the coarser grain size of large forgings or heavy rolled sections which have had less hot reduction than small forgings or small bar stock, the yield strength in the larger sections tends to be lower and the ductility not

so high as would correspond to the tensile strength of Fig. 1. This has been brought out in the general discussion of heat treatment (Vol. I, Chaps. 3 and 14). It is in the heavier sections, when both high yield

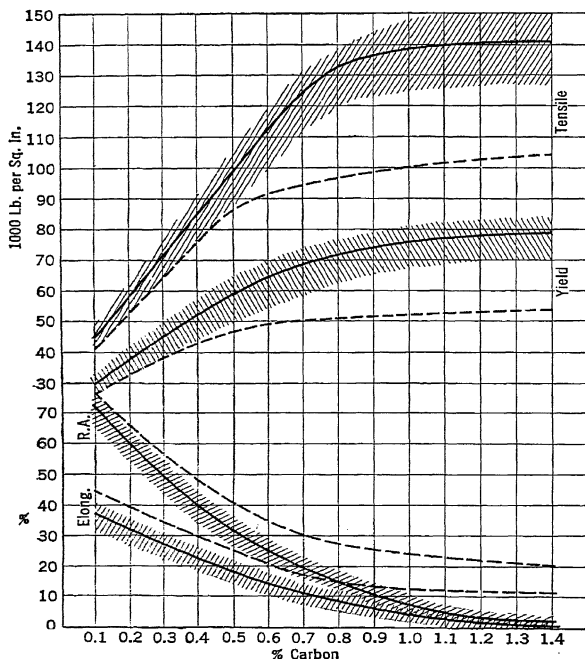


FIG. 1.—Variation in properties of normalized and annealed C steels with C content. Hatched bands, normalized steels; dashed lines, average for annealed steels. Specimens 0.505 in. diam., from bars approximately 1 in. diam.

strength and high ductility are required, that alloying is especially necessary.

**Size Effect in Steel Castings.** Data for the mechanical properties of cast steels are usually given in terms of tests on 0.505-in. bars machined from a section about 1 in. square, which has been cast as a "keel" on a heavy block designed to give ample feeding to the bar. (See Fig. 10, Chap. 1, Vol. I.) The keel is usually cut off before heat treatment, so the treatment is given to a section about 1 in. square. The properties of specimens cut from the center of large sections may

be inferior to those of the test specimen, since the casting itself may not be as adequately fed and hence may have a "loose" structure (sometimes actual shrinkage cavities), due to poor design or poor gating and risering, and, because of the slower cooling, the grain may be much larger and not as readily refined by heat treatment.

The effect of size of section in steel castings was examined by a committee of the American Foundrymen's Association \* on annealed steels of 0.23–0.27 C, 0.30–0.40 Si, 0.65–0.80% Mn. A 9-in. diam.  $\times$  18-in. high solid cylinder and the usual 1-in. keel block specimen were cast, connected. The center of the lower half of the cylindrical block compared with the standard bars as follows:

	TENSILE	YIELD	ELONG. %	R.A. %
Standard.....	68,000	42,000	32.5	53
Center.....	65,000	35,000	23	34
Standard.....	72,000	42,000	25	32
Center.....	68,500	39,000	16	17
Standard.....	65,000	39,000	27	40
Center.....	61,500	34,000	16.5	23
Standard.....	71,000	40,000	26.5	37
Center.....	64,000	37,000	12	17

Smaller variations with section size were shown by a 0.33 C, 0.30 Si, 1.00 Mn, 0.26% V steel which, double annealed from 1560° and drawn at 1380° in various sizes, gave the following:

SPECIMEN FROM	TENSILE	YIELD	ELONG. %	R.A. %
2 in. $\times$ 4 in. $\times$ 48 in.....	73,000	45,000	32	51
Center of 8 in. $\times$ 8 in. $\times$ 14 in...	70,800	38,700	33	56
Center of 15 in. $\times$ 15 in. $\times$ 15 in.	69,000	48,000	32	54
Hollow drilled from 6 in. sections of 11-ton casting.....	71,800	40,200	24	50

**Castings vs. Forgings.** The lack of hot work on the casting, as compared to a forging or a rolled section, throws all the task of refining the grain upon the heat treatment. Hence, although the tensile strength of a casting may not be far below that of a small wrought section, the yield strength, ductility, and impact resistance are likely to be lower.

\* References are at the end of each chapter. See 1.

The inclusions in cast steels are generally located at the boundaries of the original austenite grains. The inclusions are not appreciably broken up or relocated during heat treatment. Especially when the inclusions form as films rather than globules, their existence in the original locations may be a source of low ductility. This has been discussed in detail by Sims and Dahle.<sup>2</sup>

The aim in heat treatment of steel castings is to produce the best possible combination of yield strength and ductility, with the emphasis on ductility. Foundrymen have feared to use quenching and tempering, both because of the cost of such treatment and of the danger of cracking the average casting, with its changes in section, though the cast steels respond much as the wrought steels do to such treatment. Moreover, the castings are ordinarily rather massive so that quenching of plain C steels is relatively ineffective. However, especially in the alloy cast steels, oil or even water quenching is coming into use.

The older practice was to subject steel castings to a long anneal at rather high temperatures followed by very slow cooling. More recently, single normalizing, often followed by tempering, or double normalizing, first from a decidedly high temperature, followed by a second normalizing from not very far above  $A_{c3}$ , and often finally followed by tempering, has become more common than annealing.

**Silicon in Cast Steel.** It must be recalled that C steel castings are seldom exactly comparable in composition to the regular S.A.E. C wrought steels, since they commonly contain several tenths of a percent more Si, used to obtain sound castings, together with a Mn content much nearer 1% than in the usual wrought steels. These small alloying additions tend slightly to increase the tensile and yield strengths and to lower ductility slightly.

The C content of ordinary steel castings is usually 0.25 to 0.35%. Figure 2 shows the normal expectancy of small wrought S.A.E. C steel bars, taken from Fig. 1 compared with that of 1-in. keel block cast specimens.

The figures for cast steels are averaged from many sources, especially the A.F.A.-A.S.T.M. symposium.<sup>3</sup> They are above the usual specification requirements, and on the higher C steels draw temperatures up to 1300° may be necessary to attain the ductilities shown. Individual heats of C cast steel may vary quite widely from these averages.

**Cast Carbon Steels, Annealed and Normalized.** Figure 2 brings out the facts that, owing to the increased Si and Mn, the strength of cast steel is likely to be a bit above that of wrought steel of equal C

content, whereas the ductility will be lower; that normalizing and drawing tends to give both better strength and better ductility than annealing in cast material (this improvement in ductility is in contradiction to the behavior of wrought steel); and, since less strength is

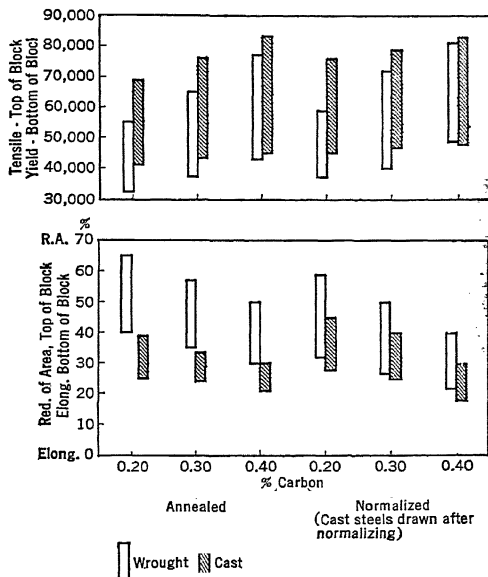


FIG. 2.—Properties of wrought vs. cast steel, in the annealed and the normalized conditions. Heat treated in about 1 in. diam. or square.

gained by increase in C in cast than in wrought steel, that around 0.20–0.30% C for cast steel combines good strength with less danger of low ductility than with higher C.

### CARBON CONSTRUCTIONAL STEELS

**Limitations of Carbon Steels.** Although technical literature devotes many more pages to the behavior of alloy steels in heat treatment, the tonnage of plain C steels, heat treated in some way, probably outranks the tonnage of all the alloy steels put together. As has been brought out in the discussion of depth hardening, Chaps. 6 and 9, Vol. I, it must be remembered that when uniform strength is desired over the whole cross section, the full effect of quenching will not pene-

trate more than  $\frac{1}{2}$  in. in any C steel, and not even that far in low C steels. Hence, when it is desired to produce a martensitic structure by quenching and to toughen this by subsequent tempering, only a partially effective job can be done on sections much more than 1 in. thick. It may be well worth while to produce a strong skin on a massive object, for the skin usually takes the maximum stress, but when very large sections or very low C steels of the plain C group are heat treated, about all that can be done is to secure increased toughness, perhaps accompanied by a slight rise in yield strength. The large increase in strength obtainable in small sections and with higher C contents cannot be obtained.

It should also be remembered that in order to secure reasonable depth hardening in any but the smaller sections of C steel, water quenching is ordinarily used, with attendant danger of warping and cracking. This trouble is decreased, but not entirely removed, by thorough homogenization before quenching or by a prior normalizing treatment.

Because the lower C steels are but slightly responsive to ordinary quenching, when similar physical properties are sought in sections in which the properties can be developed in both C steels and alloy steels, the C content of the plain C steel is higher than in the alloy steel.

**Grain Size.** It must also be recalled, as discussed under grain size, Chap. 9, Vol. I, that there are two distinct classes of C steels, those of fine grain, i.e., high coarsening temperatures, and those of coarse grain, i.e., low coarsening temperatures. The former are much more fool-proof in regard to hardening temperatures and tend to have higher impact values, though they are ordinarily shallower-hardening. Hence when data are cited for the properties belonging to steel of a given C content (the main variable), it must be remembered that the data are subject to modification in accordance with differences in grain size.

**The Size Effect.** The effect of size may be illustrated by data of the Steel Research Committee,<sup>4</sup> using a steel of 0.45 C, 0.20 Si, 0.78% Mn, water quenched from 1600°, in various sizes from  $\frac{3}{4}$ -in. to 3-in. diameter, the test specimen being cut from the *center* of the bar. It is sufficient to show the tensile strength, reduction of area, and Izod in the quenched but untempered condition and after tempering at 930°, as in Fig. 3.

The centers of the 2-in. and 3-in. bars had not been hardened by the quench; so the tensile strength at those points was not altered by the draw. However, the quench had either cooled these centers fast enough to produce a pearlitic structure that was susceptible to some improve-

ment on reheating, or else considerable internal stress was produced, which was released by tempering, since the reduction of area and the impact resistance are improved by tempering. The cooling rate in quenching at the center of the large sections approximates that obtained on normalizing much smaller sections; and, even though the structure is not visibly altered by the draw, the ductility is improved by tempering the small normalized or the large quenched sections.

**Response to Quenching.** The response to quenching of lower C steels, or to milder quenches, as in oil, is less than for the 0.45% C steel and water quench of Fig. 3. This has been quite fully discussed by Sisco.<sup>5</sup>

In very large forgings of C steel the tensile strength obtainable is primarily that due to the chemical composition of the steel, rather than to the treatment. About 65,000–75,000 for a 0.30% C steel and 85,000–95,000 for a 0.45% C steel may be obtained in 6-in. to 40-in. sections. There is no attempt to quench such large pieces of C steel, since the quenching effect would penetrate so little. Heat treatment, therefore, is confined to normalizing and tempering

or to spheroidizing annealing in order to get grain refinement due to passing through the critical in normalizing, or to develop maximum ductility and toughness. The properties given for heat-treated forging steel and for spring steel in the usual curves and tabulations based on 1-in. round bars are not at all applicable to large sections.

**Properties of Small Sections, Oil-quenched.** We shall now turn our attention to C steels in small sections, where heat treatment is of maximum effect. These will be dealt with in order of increasing C

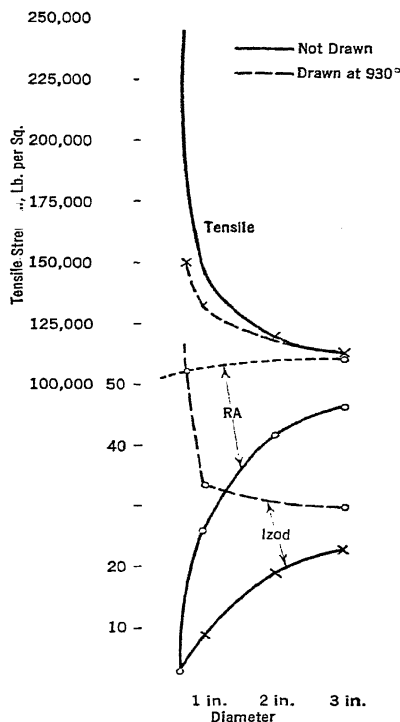


Fig. 3.—Size effect in 0.45% C steel.

Solid lines, water quenched, not drawn. Dashed lines, water quenched, drawn at 930°. Specimens taken from center of bar of size shown. (After Steel Research Committee.)



content. The effect of C is shown in the following table for specimens cut from the center of C steels of 1½-in. diam., quenched in oil and *untempered*.

%C	TENSILE
0.10	65,000
0.20	80,000
0.30	100,000
0.40	120,000
0.50	140,000
0.60	160,000
0.70	180,000
0.80	185,000
0.90	190,000
1.00	192,000
1.10	194,000

VERY LOW-CARBON STEELS, UNDER 0.15% C, SUCH AS S.A.E. 1010. Steels for vitreous enameling, tinplate, automobile body, fender, or other deep drawing stock, fall into this category. Ordinarily, though not exclusively, they are rimmed steels in which the surface is normally of slightly lower C and, at least in the ingot, of greater density and freedom from sponginess than the center. Strength is seldom a matter of concern in these services; the prime aim is ability for severe deformation. Hence the heat treatment used is for grain refinement and maximum ductility. The sheet or strip, which has been brought to the lighter gages by cold rolling, is annealed below the critical, or normalized from above it, as discussed in Chaps. 3 and 5, Vol. I. Bright annealing and bright normalizing are discussed in Chap. 12, Vol. I.

Whereas most carburizing steel runs around 0.20% C, the S.A.E. 1015 grade is held by some users to the range 0.08–0.14% C, to give maximum core softness for straightening operations, as in cam shafts. (Where this factor does not enter, the higher C 1020 is usually chosen so as to have better machinability as-rolled.) The as-rolled condition may not be satisfactory for machining with C around 0.10%, especially if the hot-rolled bars have been finished hot and so are coarse grained. The steel is so strongly ferritic that it is gummy to the tool. A better distribution of the pearlite patches may be secured by normalizing. Still better machinability of small sections can be secured by water quenching.

The very low C steels are not strongly responsive to quenching and so find but little application when the cooling rate in quenching through the critical is normal, that is, when the steel is simply immersed in a

still water or oil quenching bath. Quenching, however, favorably affects machinability. Although this very low C steel, without high Mn and P, and in either the annealed or rolled condition, does not machine freely, but is likely to tear badly in threading, fine slotting and similar operations, quenching will put the steel in a good condition for many machining operations. For such a purpose the normal treatment would be a water quench from 1700° to 1575°, according to the C content (see Fig. 7, Chap. 2, Vol. I) and mass-surface factors, with or without tempering, the tempering being omitted when conditions (of stress caused by quenching) will permit. Such a treatment will refine the grain and remove stresses set up by previous working.

The properties of a 0.10% steel may be summarized as follows:

CONDITION	TENSILE	YIELD	ELONG. %	R.A. %
As-rolled.....	51,700	35,500	34.5	65
Annealed.....	48,800	31,500	37.5	67.5
Water-quenched from 1775°...	64,700	45,500	23	61
1575° Water/1300°.....	53,000	36,500	35.5	66

The hardening effect of ordinary water quenching is also shown in the following table.<sup>6</sup>

RIVET BAR							
Steel C	0.11 C	0.01 Si	0.45 Mn	0.01 P	0.045% S		
	BRINELL	TENSILE	YIELD	ELONG.	R.A.	IZOD	CHARPY
				%	%	ft.-lb.	
As-rolled . . . . .	95	50,000	35,000	45	72	120	55
Annealed* . . . . .	85	45,000	26,000	45	72	35	9
Water-quenched† . .	135	62,000	43,000	35	72	114	60

\* Annealed at 1690–1740° and furnace cooled. † Water quenched 1690–1740°. No draw.

**Pressure Quenching.** By quenching from 1700°, a 0.12% C steel in 0.505-in. bars under a *high pressure* water spray, applied from all sides, Smith, <sup>7, 8</sup> secures 120,000 tensile, with 42% R.A. Tempering decreases strength and increases ductility, as follows:

TEMPERED AT, DEG.	TENSILE	R.A. %
400	110,000	55
600	105,000	62
800	90,000	70
1000	73,000	77

The pressure quench is commercially applied to bolts and rivets of relatively small section. Increase in quenching temperature increases the strength; 150,000 tensile has been obtained by quenching from 2300°.

To produce full strengthening, "huge volumes" of water are used in quenching under a pressure of 100 lb./in.<sup>2</sup>.

By reducing the water pressure or using a less drastic quenching medium, the tensile strength produced can be lowered and the ductility increased so that the quenched, untempered low-C specimens have about the same ductility for a given tensile as is given above for the relations at different tempering temperatures.

**Full Annealing.** In annealing 0.10% C steel it should be heated as rapidly as consistent with the size and shape of the piece to about 1650°. In these very low C steels the disappearance of the last ferrite at  $A_{c3}$  takes place very rapidly, so that it is necessary only to allow the heat to penetrate the steel at this temperature. In steels of low coarsening temperatures, in which the grain begins to coarsen rapidly with increase in temperature and length of time held there, care should be taken not to overheat and not to maintain the annealing temperature for too long a time. Cooling may be carried out comparatively rapidly without danger of hardening the steel. The annealing of these very low C steels is usually for the purpose of relieving such stresses as may have been incurred by cold work, or to efface the fibered structure of a hot-worked steel. Some grain refinement can be effected in steels coarsened in a previous process. The properties usually expected in such steel, annealed, are: Tensile, 45,000 to 55,000; yield, 28,000 to 36,000; elong., 40 to 30%; R.A., 65 to 55%.

**Grain Growth on Critical Strain in Process Annealing.** The low C steels with C up to about 0.10% are susceptible to growth of huge grains in material that has been lightly deformed, when process annealed in the neighborhood of 1200°. Material with these huge grains is brittle, and the phenomenon is termed "Stead's brittleness." The amount of deformation required to set the stage for such grain growth is critical; i.e., it extends over but a small range of deformation, the range varying somewhat with individual steels. With very tiny deformation the growth does not ordinarily occur; over the next few percent deformation it occurs readily, but at still greater deformations it ceases. The remedy is, of course, to avoid the critical range of deformation, or to soften the strained material by normalizing or by a full anneal. A small increase in C content tends toward the avoidance of grain growth. (Compare Chap. 3, Vol. I.)

**Aging Phenomena.** The precipitation phenomena involved in aging have been briefly discussed in Chap. 8, Vol. I, and have been more fully summarized by Epstein.<sup>9</sup> The effects due to the solubility relations of carbides, oxides, and nitrides are most noticeable in these low-C steels. As Herty and Daniloff<sup>10</sup> point out, these effects are

very minor in killed steels, and practically complete stability against aging can be secured by suitable "deoxidizing" additions. Griffis and co-workers<sup>11</sup> show elimination of aging by addition of 0.06% Ti and 0.04% Al, for example.

To avoid the loss from cropping off the pipe in killed steels, rimming low-C steels are used when possible, so the possibility of strain-aging and quench-aging effects is present in much steel of this particular C range.

#### 0.15-0.25% C STEEL, SUCH AS S.A.E. 1020

This grade of straight C steels, often known as "machinery steel," has innumerable uses where strength is not an all-important factor. The steel forges and machines well. In the lower C content of the range the steels find their greatest application in the case-hardening processes mentioned in Chap. 10, Vol. I, and in cold-rolled or drawn products. The higher C range is used in tie rods, nuts, flanges, pins, levers, etc.; for machine work of various descriptions; for structural purposes, in automotive construction, etc.

**Heat Treatment.** Normal quenching of steels in the lower C end of this range confers but little additional strength, except in thin sections, though small pieces, pressure-quenched, are strengthened. Quenching has a desirable influence, however, in the refinement of grain after forging or other working. Hardening must be carried out from a temperature exceeding the upper critical—which is about 1570° for 0.15% C, and about 1510° for 0.25% C—in order to effect the full absorption and diffusion of the excess ferrite. Some engineers recommend quenching the 0.15% steel at 1650° or even higher and the 0.25% C steel at 1575-1600°. Too high temperatures produce a greater tendency to warping and to oxidation, and involve higher cost of treatment. Those temperatures should be used which will produce the most efficient combination of physical properties, refinement of grain, and low cost of production. Temperatures lower than 1500° do not bring out the full effect of the treatment, as is shown by the following results upon 0.25% C steel:

Quenched in oil from, then drawn at 800°	Tensile	Yield	Elong. %	R.A. %
1450°	70,200	43,500	24	48.5
1500°	79,600	52,500	25.5	52.5

With hardening temperatures higher than 1550° for 0.20% C, there is little increase in the physical properties and, moreover, the structure may begin to coarsen rapidly. Fine-grained steels may, however, be quenched from higher temperatures. The best quenching temperature for a steel in this range of C is very dependent upon the coarsening characteristics of the steel. Water quenching of course gives high strength and lower ductility. (See Fig. 7.)

With C higher than 0.18 to 0.20%, and particularly if the section is small, or the Mn content is more than 0.60%, the necessity of drawing for toughening after quenching becomes apparent. Hardening *small sections*, such as are used in automotive construction, from about 1525° without subsequent drawing—especially if water has been used as the cooling medium—will produce a steel with insufficient ductility for most purposes. The physical characteristics of a 0.20% C steel under these conditions will be approximately as follows:

Tensile.....	90,000 to 110,000
Yield.....	60,000 to 75,000
Elong. %.....	17 to 12
R. A. %.....	30 to 15

By drawing at 800° or 900° a considerable increase in toughness and ductility is obtained, approximating:

Tensile.....	70,000 to 85,000
Yield.....	45,000 to 60,000
Elong. %.....	35 to 20
R. A. %.....	65 to 45

Cold-rolled material, subsequently given the same heat treatment as hot-rolled material of the same chemical composition, will usually show about 8,000 to 10,000 higher yield and tensile strength.

Representative values for 0.20% C steel, in small sections, are shown in Fig. 7, and characteristic results from commercial work are given in the Table on page 17.

Such values hold only for the smaller sections, below 2 in. in thickness, though some improvement is obtained in larger sections. With the increase in sectional area, the effect of hardening decreases, and for particularly heavy work it may result only in a refinement of grain. Thus, for heavy, oil-quenched forgings, toughening might not be considered a necessity; such reheating will, however, relieve the stresses which are always inherent in quenched steels and is imperative if entire freedom from warping due to internal stress is essential. Large

Material	C	Mn	Quenched in oil from °	Re- heated to °	Ten- sile	Yield	Elong. %	R.A. %
General characteristics	0.18 to 0.25	0.40 to 0.80	1500 to 1550	800 to 900	70,000 to 85,000	45,000 to 60,000	35 to 25	65 to 45
Auto. lever . . . . .	0.18	0.40	1650	800	70,030	45,400	32	64
Pressed auto. frame . . .	0.22	0.40	1530	800	71,950	43,400	29	56
Engine forging . . . . .	0.26	0.28	1650	1025	77,210	52,200	28	65
Cold-rolled $\frac{3}{4}$ in. plate . . . . .	0.24	0.60	1525	900	93,300	65,250	20.5	51

forgings thus treated will show yield strength of 30,000 to 50,000 with an elongation of 35 to 25%.

**Annealing.** There probably is more disagreement and argument as to the proper annealing temperatures for 0.15–0.25% C steel than for any other C range. Opinion and practice are divided over the use of a comparatively high temperature—50° to 100° over the upper critical—or a lower temperature lying somewhere between  $A_{c1}$  and  $A_{c3}$ . In this group the  $A_{c1}$  and  $A_{c3}$  ranges are separated widely, and the influence of the C and Mn contents rapidly increases. A high annealing temperature, 1550° to 1650° or more, will give ample opportunity for the absorption of the excess ferrite, for diffusion, and for equalization.

The question depends largely upon the initial condition and structure of the steel before annealing. If the “breaking down” during working—either rolling or forging—has been severe, if high temperatures have been used, and if the finishing temperature has not been proper, a high annealing temperature may be necessary to relieve the stresses entirely and homogenize the steel. On the other hand, if the steel has been worked carefully and the micrographic structure is good, the lower temperatures probably will be entirely satisfactory from the practical viewpoint.

If we assume as average figures for annealed steel of this C range:

Tensile . . . . .	58,000 to 65,000
Yield . . . . .	28,000 to 35,000
Elong. % . . . . .	over 30

and compare these with the results of a tensile test taken from the

steel to be annealed, a good idea of the temperature and time of heating may be obtained. For example, the following results from 1½-in. rounds for gun barrels show that a high annealing temperature was not necessary in this case, inasmuch as the original steel was in excellent condition.

Gun-barrel steel, in. rounds      P, 0.070%  
 C, 0.18%      S, 0.055%  
 Mn, 0.50%      Si, 0.055%

Treatment		Tensile	Yield	Elong. %	R.A. %
As-rolled		66,700	34,000	33.5	57.5
Annealed at	for min.				
1360–1400°	30	65,000	34,000	38.0	61.0
1500°	20	65,000	33,000	38.5	58.5
1500°	105	64,000	33,000	39.0	62.5
1830°	15	63,000	32,000	35.5	56.5
2120°	5	61,000	31,500	34	53.0

On the other hand, the following cold-rolled automobile-frame steel was particularly “hard” before annealing and required a temperature of 1550° to relieve thoroughly the effect of the cold work:

C, 0.24%      P, 0.028%  
 Mn, 0.38%      S, 0.038%

	Tensile	Yield	Elong. %
Before annealing . . . . .	100,500	68,500	18.5
After annealing at 1550° . . .	66,000	38,000	37.0

For the *average run* of annealing in this range of C, a temperature of about 1500° will be found to give satisfactory results; individual cases must be treated as such. Higher temperatures will produce structures more satisfactory for some types of service.

**Steel Castings.** When maximum ductility is required in steel castings, the C content may run below 0.25%, though the normal range in unalloyed castings is 0.25–0.35%. When the castings are to be used for their magnetic properties, i.e., “dynamo” steels, C will usually run between 0.10 and 0.15%, with Si and Mn also held as low as possible.

In general, all steel castings nowadays (unless they are used for

their magnetic properties) are given some type of heat treatment, though with steels of less than 0.20% C, in unimportant castings, heat treatment is sometimes dispensed with and the castings used "green." Since the treatment of castings of 0.20–0.25% C does not differ greatly from that for the 0.25–0.35% grade, they will be considered together in the next section.

0.25–0.35% CARBON STEEL SUCH AS S.A.E. 1030. Steel containing from 0.25 to 0.35% C is known as medium C, or soft-forging steel and is used for a wide variety of constructional purposes. It welds, forges, and machines well, and may be vastly improved by proper heat treatment. Under skillful treatment, it gives a variety of combinations of strength and ductility which probably cannot be obtained in any other range of C content.

The following results, obtained from the center of a heat-treated (oil-quenched and toughened) axle forging of 0.30 to 0.35% C are representative of the properties in large sections.

AXLE, 0.32% CARBON, ACID STEEL

Tensile.....	91,500
Yield.....	61,500
Elong. %.....	33.5
R. A. %.....	48

These steels are seldom used for case-hardening with heavy cases unless a hard, rather than a tough, core is needed to support the case. In such conditions alloy steels are normally employed, but the use of plain C, or mildly alloyed steels of this C range with very thin, gas carburized, cases, is rapidly increasing (see Chaps. 10 and 11, Vol. I).

**Untreated.** In the untreated condition, with standard Mn, P, and S, the average tensile for these steels will be about as follows:

CARBON	ACID STEEL	BASIC STEEL
0.25–0.30%	67,000 to 78,000	63,000 to 72,000
0.30–0.35%	69,000 to 83,000	75,000 to 74,000

Rolled plates from 2 to 4 in. thick, made of basic steel with 0.25 to 0.35% C and about 0.40% Mn, will usually meet the following specifications:

Tensile.....	65,000 to 75,000
Yield.....	33,000 to 37,000
Elong. %.....	30 to 25
R. A. %.....	50 to 36



These figures apply generally to as-rolled, untreated, steel of this analysis, which has had a reasonable reduction from the ingot size by forging or rolling.

In commercial rolled steel there are many gradations in composition and properties among C steels of this general range, especially in respect to the thickness of the plate. As the thickness of the plate increases, specifications generally allow a decrease in ductility and in the severity of the bend test, when that is required, from that corresponding to given tensile and yield strengths for thinner plate. Often a specification will cover two grades in the same plate thickness, in which the requirements vary only in ductility and bend test.

Fine adjustment of C content to plate thickness and the shift in strength in a given thickness with slight variation in C is indicated in the following, from current American Society of Testing Materials specifications. The specified minimum yield strength is half the tensile strength.

SPEC. NO	GRADE	PLATE THICKNESS	MAX. %C	%Mn	TENSILE RANGE	ELONGATION
A 149 :	:	under $\frac{3}{8}$ in.	0.30	0.90 max.	65,000-77,000	$\frac{1,600,000}{\text{tensile}}$ in 8-in. fire box grade
A 149 :	36	$\frac{3}{8}$ in.-2 in.	0.33	0.90 max.		$\frac{1,550,000}{\text{tensile}}$ in 8-in. flange grade
A 149 :	36	under $\frac{3}{8}$ in.	0.32	0.90 max.	70,000-82,000	$\frac{1,600,000}{\text{tensile}}$ in 8-in. fire box grade
A 149 :	36	$\frac{3}{8}$ in.-2 in.	0.35	0.90 max.		$\frac{1,550,000}{\text{tensile}}$ in 8-in. flange grade
A 70 :	36	under $\frac{3}{4}$ in.	0.25	0.80 max.	55,000-65,000	$\frac{1,550,000}{\text{tensile}}$ in 8 in.
						$\frac{1,750,000}{\text{tensile}}$ in 2 in.
A 70 :	36	$\frac{3}{4}$ in.-4 in.	0.30	0.80 max.	55,000-65,000	$\frac{1,550,000}{\text{tensile}}$ in 8 in.
						$\frac{1,750,000}{\text{tensile}}$ in 2 in.
A 150 :	:		0.35	0.50-0.90	65,000-77,000	$\frac{1,750,000}{\text{tensile}}$ in 2 in.
A 150 :	:	2 in.-4 in.	0.35	0.50-0.90	70,000-82,000	$\frac{1,750,000}{\text{tensile}}$ in 2 in.

The required elongation is decreased with increase in plate thickness by amounts stated in each specification.

\* Fire box grade. Heat treated for grain refinement, but not liquid quenched.

**Heat Treatment.** The upper critical decreases from about 1510° for 0.25% C, to about 1460° for the 0.35% C steel. A quenching temperature of 1525° for the lower C contents of this range, and 1475 to 1500° for the higher carbons will suffice under normal conditions. If the heating has been conducted uniformly and not too rapidly—especially when approaching the maximum temperature—the original structure of the steel should be entirely eliminated at these temperatures. Nevertheless, most metallurgists prefer to quench these steels from a higher temperature, say 1575 to 1600°, in order to make certain of the complete change in structure and to obtain a maximum hardening effect. The latter will be of particular value in the heavier sections wherein the rate of cooling through the critical can be increased only through the use of a high initial temperature.

For 0.35% C forgings in which especially high qualities are desired, double treatment will produce a refinement of grain and correspondingly higher yield strength and ductility than are usually obtained by the single treatment. The temperatures recommended for this range of C content are:

1. First treatment from 1600 to 1650°, using either air (normalizing) or oil quenching according to conditions;
2. Second, quenching from 1475 to 1500° followed by
3. Drawing or suitable toughening, according to the size of piece and physical properties desired.

The results obtained from heat treatment will vary considerably over this range of C. An increase of a few points in the C content is particularly noticeable in these steels. The size of the section, the quenching temperature, and the quenching medium also have large influence in this range. The yield strengths will vary from 35,000 to 80,000, with corresponding elongations of 30 to 10%. Where the maximum of uniformity in machining and in physical properties is desired, a preliminary normalizing may be used.

In quenched and tempered C steel forgings for railway axles, large shafts and the like, up to 10-in. diam. when solid, or 20-in. diam. if bored, with the specimens taken longitudinally, midway of the cross section, the minimum properties required by A.S.T.M. Specification A19-36, given below, will serve to indicate how, in spite of increase in C content, the properties fall off with increase in section. In this specification, Mn is held between 0.40 and 0.70% and quenching may be in any medium, i.e., either oil or water, as may be necessary to obtain the properties. Tempering after quenching is specified, but the temperature is not specified.

SIZE		%C	Tensile	Yield	Elong. 2-in. %	R. A. %
Diameter, in.	Wall Thickness, in.					
Up to 4	Up to 2	0.25-0.60	90,000	55,000	$\frac{2,100,000}{\text{tensile}}$ , 20.5 min.	$\frac{4,000,000}{\text{tensile}}$ , 39 min.
4 to 7	Up to 3½	0.35-0.60	85,000	50,000	$\frac{2,000,000}{\text{tensile}}$ , 20.5 min.	$\frac{3,800,000}{\text{tensile}}$ , 39 min.
7 to 10	Up to 5	0.35-0.65	85,000	50,000	$\frac{1,900,000}{\text{tensile}}$ , 19.5 min.	$\frac{3,600,000}{\text{tensile}}$ , 37 min.
10 to 20*	5 to 8	0.40-0.70	82,500	48,000	$\frac{1,800,000}{\text{tensile}}$ , 19 min.	$\frac{3,400,000}{\text{tensile}}$ , 36 min.

\*Bored.

**Annealing.** As has been previously explained, heating for annealing to just above  $A_{c1}$ , the lower critical, will refine the ground mass only, whereas complete refinement, shown by the disappearance of the ferrite network, only occurs beyond the upper critical ( $A_{c3}$ ). A temperature of 1500 to 1525° will give excellent results for the full annealing of steels within this range of carbons. Because of the air-hardening effect in small sections of steels of the higher C contents, slow cooling, either in the furnace, in lime, or in ashes, may be found necessary.

Reference should be made to Chaps. 3 and 4, Vol. I, for additional data on the subject of annealing and spheroidizing these steels.

In the annealed state, the lower carbon in this range should give

Tensile.....	60,000
Yield.....	30,000
Elong. %.....	30-40
R. A. %.....	50-60

whereas the higher carbons will usually give

Tensile.....	70,000
Yield.....	40,000
Elong. %.....	25-30
R. A. %.....	45-55

**Steel Castings.** The "regular" grade of steel castings, containing 0.22 to 0.35% C, around 0.25-0.75% Si (usually 0.25-0.60%), and 0.50-1.00% Mn, requires heat treatment to break up the coarse "as cast" structure. Even in relatively thin castings, and still more so in massive castings, it is necessary to heat castings to somewhat higher

temperatures and for decidedly longer times above the critical temperature before cooling through that temperature than is necessary with wrought steel. Heating to 1650° for 4 to 8 hr. is often required to secure satisfactory homogenization; and for large castings these may have to be exceeded.

Cooling may be carried out slowly, in the furnace—i.e., annealing—or more rapidly in air—i.e., normalizing—or in oil or water—i.e., quenching. Annealing was formerly the usual treatment, but normalizing, generally followed by a draw at 1100–1300° is now more common. On material that is homogenized with difficulty, the normalizing treatment may be repeated, usually adopting a lower temperature, say 1525°, for the second treatment. The steel need only be cooled through the critical and then immediately reheated for the second normalizing treatment.

The green casting of 0.30% C may show only about 10% elong. and R. A., with very poor impact properties. Annealing raises the ductility.

In the few cases where green castings are allowed by specifications, the specifications for ductility can only be met, without annealing or normalizing, by reducing the C, say to 0.20% or below. Green castings are covered as Grade A<sub>1</sub> by A.S.T.M. Specification A87-36 for C steel castings for railway use, but practically all the production is of grade A<sub>2</sub> or grade B, both of which must be either annealed or normalized. A draw after annealing or normalizing is optional, but if used, must be at 400° or above. Grades A and B are limited to 0.85% Mn max.; the C and Si contents are at the foundryman's option. In acid steel, 0.06% S is allowed. An alloy grade, C, is also listed, in which the foundryman may use his judgment as to C, Si, and alloying elements (among which Mn, above 0.85%, is the most used) and, with the purchaser's permission, he may cool the castings in an air blast or by a liquid quench and draw them. The required minimum properties follow:

GRADE	TENSILE	YIELD	ELONG. % IN 2 IN.	R. A. %
A <sub>1</sub> green.....	60,000	30,000	22	30
A <sub>2</sub> (norm. or ann.).....	60,000	30,000	26	38
B (norm. or ann.).....	70,000	38,000	24	36
C (alloy, norm., ann., or quenched and drawn).....	90,000	55,000	22	40

A differentiation in ductility requirements is made between castings annealed and normalized, in A.S.T.M. Tentative Specification A27: 36T for steel castings for general industrial use. This differs from

## CARBON STEELS

A37 : 36 in allowing Mn up to 1%. The specification lists the following:

GRADE	TREATMENT	TENSILE	YIELD	ELONG. % IN 2-IN.	R. A. %
A <sub>1</sub>	none	60,000	30,000	22	30
A <sub>2</sub>	norm.	60,000	30,000	26	38
A <sub>3</sub>	ann.	60,000	30,000	24	35
B	norm.	70,000	38,000	24	36
B <sub>1</sub>	ann.	66,000	33,000	22	33
B <sub>2</sub>	ann.	70,000	35,000	20	30
H	norm.	80,000	43,000	17	25
H	ann.	80,000	40,000	17	25

Castings of grade A<sub>1</sub> have to be of very low C content to meet the requirements without heat treatment. Grade H (the hard grade) requires higher C than the normal range. Ordinarily, when extra strong castings are desired, alloying will be resorted to and the properties of grade C in A87 : 36 will be sought.

**Precautions in Heat Treatment.** In the heat treatment of plain C or any other cast steels certain precautions are required. The castings should be so loaded in the heating furnace that they come up to temperature as uniformly as possible. This is accomplished by placing the castings on supports so that the furnace gases may get beneath them, instead of putting them directly on the furnace hearth or the car bottom. Heavy castings are placed on the outside of the charge, light ones in the center.

Heating time is ordinarily calculated from the time the control pyrometer shows the desired temperature (usually about 1650° for annealing or single normalizing of castings of normal composition) to have been reached, the heating-up period being lengthened for castings of heavy section. This does not ensure that the centers of heavy sections have reached that temperature, and the soaking period is increased with increase in size of section, both to secure temperature uniformity and give time for diffusion. Even the lightest castings require 1 hr. soaking, and those of 1- to 2-in. section will be given at least 2 hr. In heavier castings the usual rule is 1 hr. soak per inch of heaviest section, though 12 hr. is seldom exceeded even for the heaviest castings. It is desirable to charge large or complex castings into the heating furnace while still hot, i.e., as soon as they are shaken out from the mold. In full annealing the charge is allowed to cool to 500° or below, before removing from the furnace. If it is necessary to straighten the castings, this should be done when they are at 500–900°. The temperature is chosen as a compromise between the desirable

critical cooling rate—i.e., depth hardening and the mass effect  
rate of tempering of quenched steel  
range of properties in quenched and tempered steels, and, finally,  
as to special effects conferred by the alloying element.

The comparison of different complex steels can most readily be made by considering them in classes. These classes could be chosen on various bases; for example, in relation to the C content. Thus the low-C, low-alloy steels requiring no heat treatment; the low-C steels for carburizing, requiring carburization and treatment suited to both case and core; the medium-C, medium-alloy steels used (either annealed, normalized, or normalized and drawn) as cast steels, and almost invariably used in the quenched and tempered condition as wrought steels; the eutectoid and hypereutectoid medium and highly alloyed spring, tool, and die steels; and the hypereutectoid, highly alloyed high-speed steels fall into fairly definite classes, with the very highly alloyed austenitic steels, of whatever C content, as a separate class. By starting with the discussion of simple C and alloy steels, to bring out the proclivities traceable to each of the alloying elements, and then discussing such classes, the consideration of the more complex steels will be facilitated.

In our discussion of alloy steels we shall first take up the simple steels, those of C, and of C plus one alloying element, then consider the complex steels. However, the matters of replacement of one alloying element by another and of adding small amounts of another element to counteract the deficiencies of the single alloying element cannot conveniently be wholly postponed, so the simple steels will be dealt with on the basis of the predominant effect of the primary alloying element.

Besides the classification of simple and complex steels, a further separation into constructional and special-purpose steels is also desirable.

**Constructional Steels.** Those steels which enter into machinery with moving parts, including objects like freight cars that are themselves moved, or into structures like bridges which carry live loads, may be termed constructional steels, allowing the term to connote a need for specific strength or wear-resistant properties that may be (even though only crudely) evaluated by tensile, hardness, impact, and endurance tests. "Structural steels" is a term sometimes similarly used to include automotive and other types of constructional steels, but it is liable to misinterpretation because of the use of the term "structural steel," applied to C steel rolled into angles, I-beams, and the like for

use in building construction, thus denoting geometrical form rather than mechanical properties.

There is a certain similarity among the requirements for constructional steels, in that a proper balance of strength and toughness is sought. Moreover, such steels are only moderately alloyed, few classes having more than 5% total alloying elements. When to the requirements for some degree of strength and toughness, there is added a demand for a very particular and specialized property, like cutting ability in a tool steel, or resistance to corrosion, to oxidation, or to all the combined effects of high temperature, the steels may be very highly alloyed, they may not be heat treated at all, or if heat treated, the treatment may be aimed to develop properties of a different nature or a different order from the properties of the constructional steels.

Hence it is convenient to separate to a considerable degree the discussion of the special-purpose steels from those of the constructional type, even though some overlapping may be required.

**Wrought Carbon Steels Annealed and Normalized.** The base line for comparison of all alloy steels is of course unalloyed C steel. The strengthening effect of C in annealed and normalized steels is shown in Fig. 1. The hatched bands show scatter, which is affected by changes in Mn and Si content within the limits of "plain" C steels, by grain size, and by normalizing temperatures, whereas the line within the hatched bands shows the usual expectancy for normalized commercial steels. The dashed lines show the usual expectancy for annealed steels. Were the variations met in practice plotted as for the normalized steels, there would be scatter bands about each of these lines, analogous to those shown for normalized steels.

It will be noted that up to about 0.50% C the annealed curves follow the low side of the normalized tensile and yield bands but thereafter fall much below them, whereas the ductility curves are higher, even below 0.50% C, but deviate still more above that percentage.

The effect of cooling rate is thus most marked in the high-C steels, so that when maximum softness is wanted in them annealing must be resorted to, whereas, if strength is sought, normalizing is preferable.

It is especially to be noted that increase in C above 0.50% does not pull the yield strength up as much as it does the tensile. In fact, when yield strength is sought, alloying elements are utilized instead of raising the C, even at much lower C contents. The low yield strength (and yield ratio \*) of the C steels in the annealed or normalized conditions is one of the most potent reasons for heat treatment and for the introduction of alloys.

\* Tensile strength  $\div$  yield strength.

diffusion and homogenization that are facilitated as the temperature rises, and the undesirable grain growth that is likewise facilitated.

About 1650°, higher for very low C, lower for very high, is usually aimed at with steel that coarsens readily and is not to be re-treated. By use of suitable grain-control additions, grain growth can be avoided, or it may be by a subsequent normalizing, repaired, and higher temperatures with either shorter times if they will bring about homogeneity, or still longer times if they are needed, can be employed.

**Normalizing.** When a normalizing treatment is employed, after a full anneal, or more commonly, after a preliminary normalizing, heating for this second normalizing should be only just high enough and just long enough for the casting to become completely austenitic throughout; i.e.,  $A_{c3}$  is not greatly exceeded, and 1525° is customary for steel of normal composition. Diffusion has been accomplished in the prior treatment, and the aim now is to secure refinement without allowing any opportunity for grain growth. This second treatment may often be avoided by use of a grain-size control addition that prevents grain growth in the first heating for normalizing. Practice is trending toward the use of compositions that will respond satisfactorily to single treatment.

**Tempering.** Tempering is seldom required after full annealing, but is quite generally applied after either a single or the second of a double normalizing. Tempering may be merely for stress relief, say 2 hr. at 750°, which does not affect the mechanical properties, or at 1000-1300°, to lower strength and improve ductility. Specifications seldom require tempering; tempering is employed to secure the required ductility when this has not been obtained in prior treatment, or to improve the impact. Hence it is ordinarily permissible to select a composition that will meet the specification merely by a single normalizing, without the expense of tempering. Slightly alloyed, grain-size controlled compositions are therefore growing in favor.

**Normalizing vs. Annealing.** No difficulty is met in meeting the tensile and yield values of grade B, but to produce them together with the required ductility, requires steel that has been correctly melted and deoxidized. If it has been, so that the steel correctly responds to heat treatment, normalizing of small sections will tend to give better than 75,000 tensile and 45,000, or higher, yield, with ductility figures usually better than upon annealing. The better tensile and yield for the same or better ductility argue in favor of normalizing. The saving in time and fuel of air cooling *vs.* furnace cooling is offset if it is necessary to temper after normalizing in order to relieve cooling strains and develop the best ductility. For extraordinarily heavy castings, nor-



malizing from 1650°, annealing from 1550°, and a long spheroidizing soak at 1250-1300° may be required to break up the cast structure and develop maximum ductility.

**Quenching.** Although no question of transformation to martensite occurs in the air cooling of massive castings, of steel of this C content with normal Mn, the type of pearlite formed on cooling through Ar<sub>1</sub> is affected by differences in cooling rates in the furnace and in air.

Increasing the cooling rate by water quenching of sections around 1¼ in. thick, followed by tempering, gives still better mechanical properties. Ninety thousand tensile, 60,000 yield, over 25% elong., and over 50% R. A. may be obtained by quenching thoroughly homogenized castings in water, then tempering at 1150°. Despite the fear of cracking complicated castings held by many foundrymen, well-designed, well-filleted, sound castings, thoroughly homogenized before quenching, are being quenched commercially without cracking difficulties.

In heating for quenching the castings will ordinarily have been previously annealed or normalized (since such treatment for homogenization diminishes the danger of cracking), so the temperature for the quenching operation is usually lower than for annealing or normalizing, and the heating time is shorter. Instead of 1 hr. soaking for every inch of thickness, ¼ hr. is customary, and the temperature ranges from around 1625° at the low-C end to 1575° at the high-C end of the customary C range. Cold water (about 70°) is ordinarily used for steel at the lower C end of the range, and either oil or warm water (about 125°) for those at the higher end. The quenched castings must always be tempered. Tempering temperatures range from 800 to 1300°, with the time around 2 hr. at temperature. Especially if the C is on the high side, tempering should be carried out immediately before the castings get cold. Although some very intricate castings are satisfactorily quenched, those that have both heavy and light sections are very prone to crack.

Differential quenching, i.e., local quenching of certain parts of the casting by a dip or a spray, with the rest of the casting protected from the coolant, is sometimes applied. Such an operation requires very great care and precise control.

The properties obtained by varying the draw temperature of a quenched 0.30% C cast steel fall closely on the curves usually given for quenched and tempered, wrought S.A.E. 1030 steel, in similar small sizes. The effectiveness of quenching naturally decreases as the size of the casting increases, so that in very massive castings quenching is

little more effective than normalizing. In heavy castings the grade C properties are therefore sought by alloying and normalizing.

For an extended summary of the properties of heat-treated C steel castings, the reader is referred to Chaps. 2 and 3 of Sisco<sup>5</sup> and to Lorig and Williams<sup>12</sup>. These authors include scattered data on castings of 0.40-0.60%, and even higher, cast C steels. These find a little use where the abrasion-resisting properties of higher C steels are required, but the tensile and yield strengths are, in general, but little improved over those of the regular grade, whereas the ductility and impact resistance of annealed steels are sharply reduced. To bring out their best properties requires quenching and tempering. When quenching and tempering are applied to high-C cast steel, oil quenching is favored to avoid quenching cracks, though hot water finds some use. By tempering the quenched 0.40-0.50% C steels at 1250-1300°, tensile of 85,000-90,000, yield of 50,000, with 25% elong., 40% R. A. can be obtained, whereas normalizing and drawing tend to produce less ductility, especially as to R. A., and the annealed steels show still less ductility and lower yield strength.

Since it takes quenching to develop the best properties of the high-C cast steels, and the propensity toward cracking increases with the C, production of steel castings over 0.35% C is very limited.

**Higher Carbon Wrought Steels.** Although 0.40% C is "high" C for castings, it is a normal content for wrought plain C steel of forging grade which is to be heat treated. The coarse structure of the casting or ingot is broken down in hot forging or hot rolling, so that the steel is amenable to heat treatment, giving, in small sizes, after quenching and tempering, ductilities scarcely inferior at a given tensile to those of wrought steels of the next lower C range. Since higher tempering temperatures are required to soften to the same degree, quenching stresses are even more satisfactorily removed.

At this range of C content, oil quenching of small sections is almost as effective as water quenching, so that the reduced warping and smaller danger of cracking favor the less drastic quench.

The C steels are still in the "water-hardening" class up to perhaps 0.60% (and, of course, water quenching is sometimes used with still higher C when hardness rather than toughness is sought, as in tools) but at about 0.40% C it becomes necessary to compare the advantages and disadvantages of the water and oil quenches.

#### 0.35-0.45% CARBON STEEL SUCH AS S.A.E. 1040

Straight C steels with 0.35 to 0.45% C—"forging steels"—are particularly suited to medium and heavy forgings for which the lower-C

steels will not give sufficient strength. Steel of this analysis is commonly used for high-duty and moving machine parts; for axles, side-rods, crankpins, and other locomotive forgings; for guns and gun forgings; for crankshafts, driving shafts, and similar engine parts; and for general structural and engineering purposes requiring a combination of high strength and good ductility. These steels are readily machinable in the annealed or soft-toughened condition, but should not be used for screw machine stock.

**Heat Treating Small Sections.** The criticals are approximately 1470 and 1430° for 0.35 and 0.45% C steel, respectively, with normal Mn content. Therefore the average hardening temperature for small sections will be in the vicinity of 1525 to 1475° for steels whose grain is readily coarsened. The fine-grained steels will stand somewhat higher hardening temperatures. Either oil or water may be used as the cooling medium, but in almost all cases hardening must be followed by a toughening operation to remove the strains, impart the necessary degree of ductility, and permit such machining operations as may be necessary.

The results obtained from water and oil quenching and subsequent toughening of small rounds of 0.35 and 0.45% C steels are given in the charts of Figs. 8 and 9.

**Heat Treatment of Heavy Sections.** From the economic standpoint, the heat treatment of heavy sections is predicated on the basis that only those forgings will be heat treated (quenched and toughened) in which heat treatment will yield returns, in increased physical properties (strength and ductility), commensurate with its cost. Mere mass in a crankshaft does not necessarily signify that the design and construction are sound economically. Many Diesel-engine crankshafts, forged and heat treated under expert supervision, have actually cost less, through the savings effected in the amount of initial steel and forging costs; and the higher physical properties and lighter weights have permitted many refinements in design. Thus, sections of, say, 5- to 15-in diam., showing 30,000 to 40,000 yield in the annealed or forged condition, may be raised to 50,000 to 60,000 by suitable forging and heat treatment. On the other hand, if the cross-sectional area is of such magnitude that the heat treatment will not pay for itself in terms of lower initial cost or better or lighter design, then heat treatment is not warranted economically.

Although the heat treatment of heavy sections is an art in itself, no principles other than those indicated in Vol. I. are involved. There must be a definite and satisfactory correlation of (a) initial structure,

## HEAVY FOR

(b) temperature and time of saturation, (c) heating for toughening.

The initial structure, as presented for heat treatment, is dependent largely upon the heating for forging, the temperature of finishing, the amount of mechanical working and reduction from the ingot, and the quality of the steel itself as defined by the melting and casting methods. Unless these factors bear a proper relation to one another, the metallurgist will soon find that his real job is an attempt to put back into the steel that combination of qualities which has previously been destroyed by improper methods. His department, therefore, becomes a hospital for sick steel instead of the scene of a manufacturing process. Because of the vital importance, common occurrence, and economic wastefulness of such conditions, Chap. 14, "Proper Heating in Forging and Rolling Practice," has been included in Vol. I.

Assuming, however, the existence of the undesirable initial structure, the metallurgist is faced with the necessity of overcoming the handicap presented. In Chap. 3, Vol. I, an attempt was made to classify the initial structures thus to be found, and to indicate methods of heating which will tend to "bring back" the steel to a normal structure. The first part of Chap. 6, Vol. I, "Principles of Hardening," also developed the same proposition. Both discussions refer especially to this heat treatment of heavy forgings and therefore apply with particular emphasis to the 0.35 to 0.45% C steels under discussion. It will thus be evident that homogenization has a vast influence and that the heating phase is the preparation for the cooling cycle.

With the establishment of a proper initial structure, the cooling cycle determines the final physical properties. This involves a consideration of the factors of maximum temperature, size of section, and cooling medium; and these establish the cooling rate. In order to obtain, as nearly as possible, the critical rate of cooling in the mass as a whole, the maximum temperature must be increased as the section increases, but with the limitations evoked by the development of grain growth. The cooling rate is increased further by using cold water with as large a volume velocity as local conditions will permit. For crankshaft forgings having journals 8 to 12 in. in diameter and corresponding throw-sections up to 12 by 18 in., an average hardening temperature of 1550° with water quenching has proved satisfactory. As in all such cases, when fine-grained (high coarsening temperature) steels are employed, the quenching temperature may be raised. If a preliminary high temperature homogenizing treatment, to be followed by either air, oil, or water quenching, is used, a temperature of 1750° is recommended for that operation. Contrary to the opinions held by many,

drastic water quenching of heavy forgings may be safely used, *provided* the steel is clean and the forgings are *thoroughly* homogenized at the recommended temperatures. An immediate toughening to the required temperature—generally between 1050 and 1200°—is always advisable.

Heat treatment of heavy sections may generally be adjusted to give tensile, 80,000; yield, 50,000; elong. 22% R.A. 45%.

Somewhat smaller sections, as in locomotive axles, may be quenched from a lower temperature—1450°—in oil, to give similar properties. In one plant the following toughening temperatures, in relation to C content, were used:

	TEMPERATURE
0.42 to 0.45% C.....	1175°
0.38 to 0.42% C.....	1125°
0.33 to 0.38% C.....	1075°
0.28 to 0.33% C.....	1000°

#### 0.45–0.60% CARBON STEEL SUCH AS S.A.E. 1050

**Treatment of Large Sections.** As the C content is increased progressively beyond 0.45%, its effect becomes quite noticeable in the added “brittleness” of the steel. Hence there is little quenching of large sections when the C content exceeds 0.50%. The danger of cracking in the treatment itself and of possible fracture in service almost prohibits such treatment of large sections. Thus the full heat treatment of large sections even though it may bring out higher physical characteristics of the steel, as is shown by the figures below, obtained from the treatment of a 0.50% C axle, is not an important factor in steels of such C contents.

	Forged	Quenched in Water from 1400° Toughened at 1200°
Tensile.....	83,500	90,000
Yield.....	34,500	53,500
Elong. %.....	22.5	27.0
R.A. %.....	30.0	53.0

**Tempering of Small Sections.** On the other hand, the hardening and *tempering* (as distinguished from *toughening* or drawing) of the smaller sections, such as gears, dies, etc., begins to take an important place in heat treatment of these carbons. In such cases the increased C content brings about an inherent hardness for resistance to wear, which is developed by hardening and tempering. The medium-sized

and smaller sections may be hardened satisfactorily in water with but a small proportion of the danger which would result from the water quenching (or even oil quenching) of larger sections. By varying the reheating temperatures, the following results may be obtained:

Yield.....	50,000 to 110,000
Elong. %.....	20 to 5
R.A. %.....	50 to 15

**Annealing.** The commercial annealing of steel, 0.50 to 0.60% C, may give a variety of results. This is due largely to the effect of different rates of cooling in relation to the size or mass of the steel. To illustrate: 6 by 6-in. billets of 0.50 to 0.55% C which have been heated to 1400° and furnace-cooled will, in general, show:

Tensile.....	80,000
Yield.....	40,000
Elong. %.....	22
R.A. %.....	35

On the other hand, smaller sections annealed in the same manner and in the same furnace, will, according to their size, give results of:

Yield.....	45,000 to 60,000
Elong. %.....	20 to 15
R.A. %.....	40 to 30

The extreme variability in the rate of cooling, as dependent upon the size of section and mass of the steel, its relation to the size of the furnace, the degree to which the cooling of the furnace may be controlled, and numerous other related factors, makes the commercial annealing of these steels an individual problem as far as actual physical results are concerned.

It is always advisable, therefore, if specific results must be obtained by annealing (used in the broad interpretation of the term), to take first a preliminary test of the steel in the condition in which it is received. From such results it will then be evident how much the steel must be "let down," and the proper reheating temperature may be judged from previous experience or by experiment. Although annealing at a temperature under the critical range will not change the general structure of a pearlitic steel, it will relieve the stresses, and thereby improve the steel. However, the previous working, such as rolling or forging, which the steel has undergone, will, in a majority of

cases in actual practice, leave the steel in a condition such that a reheating—or commercial annealing—will change the physical results, even though the annealing temperature is *under* the critical range.

**Carbon Steels in the Eutectoid Range.** Although, as was pointed out above, quenching and tempering operations are handicapped with steel much above 0.50% C, there are important applications for heat treatment on steels in the range between 0.50% C and the hyper-eutectoid compositions that contain the free cementite needed for wear resistance in the tool steels. Carbon steels, approximating eutectoid composition, are employed for rails and even higher carbons for springs.

Small springs, less than  $\frac{3}{16}$  in. thick for flat springs or  $\frac{1}{2}$ -in. rod diam. for coiled springs, may be formed cold from cold-drawn "patented" wire (see Chap. 5, Vol. I) or from "hard-drawn" wire, similarly cold drawn but process annealed between passes instead of patented. The forming or coiling is followed by a stress-relieving draw, usually at 400 to 700°. The hardness of the spring is ordinarily held in the range 375–450 Brinell. Similar hardness may be obtained by drawing to size, annealing, oil or lead quenching, and tempering. Such so-called "oil-tempered" spring steel is also used for cold forming in the small sizes. In the use of any of these grades, only the stress-relieving draw is used after forming.

An alternative is to cold form in the small sizes, or hot form in larger sizes, using annealed stock that has received a spheroidizing treatment. (It is highly important that the annealed stock be free from decarburized skin.) The formed spring is then oil quenched and tempered, again to the usual 375–450 Brinell range. Heating for quenching should also be so conducted that the stock is free from decarburization.

The C steels used for small springs may be relatively low in C, 0.45–0.70%, but high in Mn, around 1%, or higher in C, 0.80–1.05%, with normal Mn. In annealed stock to be quenched and tempered after hot forming, fine-grained steel of 0.85–1.05% C and 0.25–0.50% Mn is ordinarily specified. The hot forming of the heavier springs, performed at around 1450–1600° and preferably near the lower limit, to avoid grain growth and to minimize decarburization, serves as a normalizing operation. Very heavy springs are formed at higher temperature, about 1700°. The springs are reheated to 1450–1550°, oil quenched, and tempered at 700–1000° to the usual range of spring hardness. Effective hardening is limited to rod diameters under  $1\frac{3}{4}$  in., or equivalent flat sections. Because of the importance of avoiding decarburization, lead bath heating is often preferred.

**Hypereutectoid Carbon Steels.** A 1% C steel in, say, 1-in. bar stock, quenched and drawn back to spring temper, has a tensile of the order of 200,000, a yield of 125,000–140,000, with 10–12% elong., and 25–35% R. A. Increasing the tempering temperature to 1250° drops the tensile to around 125,000, the yield to around 90,000, and raises the elongation to 20–25%, and the R. A. to 45–50%. As has been discussed under “austempering,” high-C steels tend to have “micro-cracks” in the martensite formed in quenching so that the ductility and impact properties obtained when such cracked martensite is tempered are more likely to be low and erratic than when an equivalent set of properties is obtained with a medium-C steel. Hence, as structural materials, the high-C steels find little use. Their field is where the service demands hardness and strength rather than toughness.

**Notch Sensitivity of High-C Steels.** In “spring temper” practically all the steels are notch-sensitive, the yield strength being so high that at the tip of a notch, or of an advancing crack with its high stress concentration, the metal does not yield over any appreciable volume to distribute the stress; so the crack advances rapidly. The stress-distributing quality of the softer steels as compared to its lack in the harder steels is somewhat analogous to boring a good-sized hole ahead of a crack in a plate-glass window. When the crack reaches the hole, the stress is distributed over a considerable volume instead of being concentrated, the stress concentration dies out, and the crack stops. Poor fillets, surface imperfections, or non-metallic inclusions in the steel itself, all of which act as stress raisers, are, therefore, far more fatal in hard steels than in soft ones. (Compare Fig. 15, Chap. 1, Vol. I.)

**Inclusions.** An exhaustive study by an A.S.T.M. committee on S in rivet steel showed that MnS inclusions had strikingly little effect upon any property, including endurance, of 0.10% C steel; and the low-C free-machining screw stock in which S and MnS are sometimes raised to ten times the normal content is likewise but very little affected in static properties. But the harmlessness of stress raisers shifts into harmfulness when the steel is made very hard. This, plus the difficulty of machining, is the fundamental reason why steel of 200,000 tensile is not ordinarily selected for structural uses.

**Carbon Steel Wire.** There has been one very interesting attempt to use steel of this general type for a specific structural use, which deserves mention because of the wide publicity that was given to the troubles resulting from the attempt to use quenched and tempered high-C wire for bridge suspension cables.

Patented cold-drawn wire which has been processed by passing



the wire through a lead bath at about 1850°, thence to another lead bath at about 900°, producing a fine primary troosto-pearlitic structure of about 400 Brinell without the intermediate formation of martensite, followed by cold drawing, can be given the following properties by drawing to 0.192 in. wire: 220,000 tensile, 183,000 yield, 84,000 proportional limit, 6.5% elong. in 10 in., with a R. A. of 33 to 42%, average 40.

Instead of using such wire in the cables of the Mt. Hope Bridge at Providence, R. I., and the Ambassador Bridge at Detroit, Mich., heat treatment alone without cold drawing was resorted to. The steel contained about 0.75 C, 0.50 Mn, 0.15 Si, .02 P, and 0.03% S. The wire, drawn to the final size of 0.192 in., was rapidly passed through a lead bath at 1330° (24 sec.) for preheating, then into one at 1550° (35 sec.), then into one at 1250° (33 sec.) (thus hot quenching the wire). The wire was then passed into another 1550° bath for 37 sec., from which it was passed through oil at about 250° for 90 sec., through a sand box for 10 sec., and tempered in another lead bath at about 830° for 18 sec. The wire was then pickled, washed, fluxed, and hot galvanized in molten zinc at 850°.

If we examine the "austempering chart," Fig. 10, Chap. 6, Vol. I, for a rather analogous steel, we see that in the oil quench at 250°, the transformation of the austenite starts in 10 sec., is only about one-fourth completed in 100 sec., but is not entirely completed in 1 hr. Some austenite must thus have been retained after emerging from the oil quench so that further cooling during passage through the sand may have allowed the formation of some martensite, which, if formed, would be broken down in the subsequent tempering. It is not clear whether the heat-treatment cycle fully avoided the formation of martensite needles that might have had micro-cracks in them.

However, the finished wire had the following properties: 220,000 tensile, 190,000 yield, 100,000 proportional limit, 6% elong. in 10 in., and an average of 37% R. A. The improved yield strength and proportional limit indicated, offhand, that the heat-treated wire would be superior.

At an advanced stage in the erection of the bridges, wires in the cables began to snap. So many broke that repair was impossible, and the cables had to be removed and replaced by others made up from cold-drawn wire processed in the usual manner.

An exhaustive study of the causes of failure was made by the Bureau of Standards.<sup>13</sup> In this study it was found that *sound* heat-treated wire acted as it was supposed to from its higher proportional limit. When loaded for 48 hr. at 150,000 lb./in.<sup>2</sup>, i.e., above the pro-

portional limit and below the yield, the heat-treated wire stretched but 3%, while the cold-drawn wire stretched 12%. It was found also that the surface of the wire had occasional cracks, extending in fairly deep, with sharp ends, that were filled with brittle Fe-Zn alloy. When tensile tests were made on such fissured specimens, the reduction of area was low, as low as 7%. Non-fissured specimens gave values as high as 53%. When the Zn and Fe-Zn alloy layers were removed and the steel itself reduced in diameter enough to "clean up" below the alloy-filled fissures, no low values were found.

**Fatigue of Wires.** Fatigue tests, made on heat-treated and cold-drawn wires, both galvanized, or both stripped, showed little difference between them, until tests were made over the stress range 600 to 25,000 lb./in.<sup>2</sup>, with the wires looped over a pulley of radius corresponding to that of the anchor shoes over which the ends of the wires were looped in the cables themselves. Under such a test, with the wires pre-formed (bent) as they were in the bridge construction, the heat-treated wires failed in less than a million cycles, while the cold-drawn wires did not break. It was found that cold-drawn wires, thus pre-formed, fitted the shoe smoothly, i.e., had taken a permanent set due to the lower proportional limit; but the heat-treated ones did not, opening up, after removal, to a wider radius of curvature, and often to an irregular curve instead of a smooth one. It was then deduced that this failure to fit the shoe caused the introduction of bending (not tensile) stresses under fluctuating wind loads on the cable and its resultant swaying, which were of such magnitude as to exceed the endurance of the heat-treated wire under those conditions.

The fatigue tests on stripped wire made it appear that, in the absence of the brittle Fe-Zn alloy layer, either type of wire should have had sufficient fatigue resistance and indicated that the alloy-filled fissures of the heat-treated wire were a primary cause of the trouble, the failure of the heat-treated wire to seat itself on the anchor shoe being another primary cause. That the hot-galvanized, cold-drawn wire having the brittle alloy layer (but no fissures, since the drawing of the wire through the die had compacted and smoothed its surface) did not fail indicated that the fissures made the difference. The heat-treated wire showed a much rougher surface than the cold-drawn wire after chemically stripping the Zn.

The inability of steel of so high tensile and yield strengths to resist the effect of surface stress-raisers was thus demonstrated in dramatic fashion.

**Rails.** American rail steel used for heavy rails contains 0.69-0.82 C, 0.70-1.00% Mn, and so approximates eutectoid composition.

Until very recently, all rails were used in the as-rolled condition. The physical properties, as-rolled, run about 140,000 tensile, 85,000 yield, 9% elong., 12% R. A., 300 Brinell, and an Izod impact notched-bar value of only about 2 ft.-lb. The impact resistance is so low that unnotched bars are generally used to trace small differences in behavior. Unnotched Izod bars at room temperature give 20-40 ft.-lb. At low temperatures the ductility and impact resistance fall. Wishart and Boone<sup>14</sup> show that at  $-50^{\circ}$ , elongation and reduction of area may drop to 3% each, and unnotched Izod to 2 ft.-lb.

**Shatter Cracks and Transverse Fissures.** The rail has an irregular shape and, in order that it may be straight after cooling, it is rolled with a camber, coming from the rolls curved but straightening as it cools. This is visible evidence that cooling stresses are present. Some heats of rail steel are unable to yield to these stresses, and during the cooling process tiny internal cracks called "shatter cracks" may occur, chiefly oriented at right angles to the rail length. (See Fig. 15, Chap. 1, Vol. I.) There is no external evidence of the presence of such cracks, and shatter-cracked rails will pass all the regular acceptance tests.

Under the variable loading on the rail, conditions are set up for fatigue failure of rails containing internal stress-raisers such as shatter cracks. It is now universally accepted that under repeated loading a shatter crack may act as an internal nucleus for a spreading failure (such as would be expected in these hard steels on the basis of knowledge of their notch sensitivity), which extends transversely through the head of the rail and may thus be almost entirely cut through from the inside with no external evidence. In recent years the use of the Sperry "detector car," which passes a heavy electric current through the rail in the track and magnetically locates and records the impediment to the current that exists at the location of a transverse fissure, has become common. This non-destructive method of testing has allowed the removal of fissured rails (and of other rails of the same heat) in time to prevent many accidents that would otherwise have occurred. Moreover, the examination of fissured rails removed from service has strengthened the belief that there is a causal connection between shatter cracks and fissures. Unfortunately, the Sperry method is not sensitive enough to detect shatter cracks, so that rails with shatter cracks cannot be sorted out and scrapped at the mill. The rail must be in service long enough for a fissure to develop to about the size of a dime before its presence can be noted.

**Retarded Cooling.** Hence attention has been directed toward some treatment that will assuredly prevent shatter cracks. In normal operation, only one heat of rail steel out of 50 shows the cracks in regular

as-rolled material, but that is too large a proportion for railway safety. The difference between the one and the 49 has not yet been clearly enough demonstrated \* so that the one can be avoided by other means with sufficient certainty, and the only method applicable at the moment is to so heat treat all heats that shatter cracks are prevented.

The heat-treatment method adopted is to retard cooling during the temperature range in which the steel is otherwise too stiff to yield under rapidly applied cooling stresses without cracking. Thus time is given for yielding, and the ability of the steel to "creep" is invoked.

This temperature is a fairly low one. There is some argument as to just how low it is, i.e., at what temperature one may cease to delay the cooling. Some say 650°, others 400°. The retarded cooling does not have to be started above 950°.

The "controlled-cooling" treatment then consists in transferring the rail from the hot bed, before it has cooled below 950°, to a well-insulated chamber in which it, with its fellows, remains till it has cooled so slowly to the temperature at which shatter cracks no longer occur that cooling stresses are not built up above the capacity of the steel to withstand them. So far, the controlled-cooled rails have shown an encouraging freedom from shatter cracks or, in service, from transverse fissures.

**Rail Wear.** Transverse fissures are not the only factors that shorten the service life of rails. Wear of the rail head and "batter" of the rail ends, i.e., peening down of the rail end as the wheel drops on it as it passes the rail joint, are troublesome from the economic, though not from the safety, point of view. Obviously, the way to combat general head wear and end batter is to make, in one case, the whole head, in the other, the ends, harder.

**Heat Treatment of Rails.** Quenching and tempering would, of course, accomplish this. In short lengths and in some special cross-over constructions, the whole rail is oil quenched and tempered by lowering the rail endwise into the quenching bath. Goodaire<sup>16</sup> reports 165,000 tensile, 125,000 yield, 360 Brinell, 14% elong., 49% R. A. for

\* There is rather impressive evidence that shatter cracks in rails and probably "flakes" in some alloy steels are connected with the presence of very small amounts of hydrogen,<sup>15</sup> introduced during melting and not given off when the steel freezes. If this be correct, avoidance of the introduction of hydrogen in melting might prevent the whole difficulty. Slow cooling may allow the escape of hydrogen before the steel cools to a temperature where it is insufficiently plastic to withstand the combination of cooling stress and internal pressure of hydrogen. Until the correctness of the hydrogen hypothesis becomes more evident and sure means of avoiding introduction of hydrogen are worked out, slow cooling is the logical solution of the problem.

the completely quenched and tempered rail, which is a marked improvement in ductility over the ordinary rail.

The quenching endwise of 39-ft. rails would be inconvenient, if not commercially impossible. Immersing the whole rail sidewise in oil or water and letting it harden throughout would result in a rail so warped and twisted as to be useless. However, it is only necessary to harden the rail head to a small depth. Momentary quenching of the surface, then allowing the heat in the body of the rail to draw the surface back, is being applied to a limited extent. In one (the Kenney<sup>17</sup>) process the rail, heated above the critical, is dipped, head down, into water for 30 sec., lifted out, at once transferred to an equalizing furnace held at about 1000°, and then slow cooled, as in the regular slow-cooling procedure, to avoid shatter cracks. In another, the Sandberg<sup>18</sup>, a definite amount of water in a very fine mist is sprayed upon the rail head and the rail then transferred to the slow-cooling chamber.

In another<sup>19</sup> the rail, as it leaves the hot bed, is put into a continuous furnace 250 ft. long, before it has cooled far below the critical, is there reheated above the critical and the temperature equalized. The rail is then cooled rapidly enough through the critical in the next section of the furnace to give a normalizing, grain-refining, effect, and finally carried into the slow-cooling zone to retard cooling through the range of "shatter temperatures."

By these various means the surface of the rail head, if made martensitic, is immediately tempered. The steel a little farther beneath the surface is cooled at a sufficient rate to produce fine pearlite or primary troostite-pearlite by a sort of "austempering" process. In the water-quenching processes the rail head is brought up to about 360 Brinell, the tensile to 180,000, the yield to 125,000. The ductility figures are not appreciably altered, but the impact resistance is somewhat raised.

In the normalizing process, followed by controlled cooling, the tensile and yield are not affected, or slightly lowered, and the hardness slightly reduced; but ductility is improved, elongation of 17% and reduction of area of 27% being reported. Low-temperature impact resistance (as measured by unnotched Charpy tests on large specimens) is also materially improved. At -50°, where this particular unnotched specimen on the usual rail or the unnormalized controlled cooled rail shows 10 and sometimes as low as 2 ft.-lb., the normalized rail may give 100.

**End Hardening.** Since end batter is even more troublesome than is wear over the main length of the rail, some mills are hardening the

rail ends more drastically than by the mild methods mentioned above, by applying a water jet, for a definitely controlled period, to the ends while the rail is above the critical, and allowing the heat of the rail to draw it back to 400 Brinell. An analogous end-hardening process may be applied in the course of the continuous normalizing process. With the rail above the critical, a strong air blast is blown upon the hot rail head at the ends, which is thus locally hardened to about 360 Brinell.

Rails in service that have not been end hardened at the mill are sometimes heated at the end by a blow torch until the zone to be hardened is at a temperature above the critical, quenched by a water spray (with a definite quantity of water), and drawn by the stored heat of the preheated rail, or by a smaller flame. Naturally there is more danger of spalling when treating a cold rail in this fashion than in hardening the end of a hot rail in the mill.

Battered rail ends in the track are also built up by welding, the rapid cooling of the deposited weld metal by the mass of metal to which it is attached producing a sort of quench. The deposit is peened and ground down to the proper level.<sup>20</sup>

Bronson<sup>21</sup> doubts the value of end hardening by water or air quenching either in the mill or in the track, feeling that the control of heat-treatment conditions is not sufficiently precise to avoid danger of a poor transition zone. The American Railway Engineering Association,<sup>22</sup> however, evidently considers it a coming practice, since experiments are being carried on with all the methods for surface hardening at the ends. The depth of hardening used on rail ends is very much greater than that sought in ordinary surface hardening such as was discussed in Chap. 13, Vol. I.

Tool steels are an important group of high-C steels which will be discussed in Chap. 21.

**Mechanical Property Charts—General Discussion.** There have appeared in the "Handbook of the Society of Automotive Engineers" for many years back, mechanical property charts for most of the S.A.E. steels, quenched and tempered in small sizes, some of which have been copied in the "Metals Handbook" of the American Society for Metals, and in previous editions of this book. A warning was always given that the values should not be used for specification purposes, that they were not applicable outside the range of  $\frac{1}{2}$ - to  $1\frac{1}{2}$ -in. diam. bars and that they were "conservative" values. By "conservative" was meant that values somewhat lower than the average, or the expected, values, were given for *all* the properties.

Since strength and ductility change in inverse ratio with tempera-

ture, steel with somewhat lower than the average strength will ordinarily have somewhat above the average ductility, so such plots on the "conservative" basis are innately inaccurate.

Moreover, an S.A.E. number allows a tolerance in C content, ordinarily of 10 points ( $\pm 0.05\%$  C) and usually a considerably larger tolerance in content of each alloying element. If steels at the upper and the lower ends of the tolerance range are compared, the results will be materially different.

Again, coarse-grained steels harden to a greater depth than fine-grained ones, and the former ordinarily show slightly better ductility and materially better impact for a given hardness or tensile than the latter (Chap. 9, Vol. I).

Any one bar of steel, quenched and tempered at different temperatures, should show quite smooth property curves, but two different lots of steel, each well within the composition tolerance for the S.A.E. number, may vary considerably. If grain-size variations are superimposed on composition variations, the spread may be further widened.

Reference to the "probability" curves for S.A.E. 6130 and 3130, Figs. 24-29, Chap. 1, Vol. I, will show that property curves should be plotted as bands rather than as lines. The amount of work necessary to establish the width of the bands that would truly represent the variability in commercial steels of even a single S.A.E. number is huge, and for all S.A.E. and other commercial steels, is enormous.

When any property chart is inspected, the reader should visualize a band about each line, and should remember that the line shown is not necessarily in the middle of the band; the band might lie mostly above or below it. In the old S.A.E. property charts the band should lie mostly above the line. The curves put out by conservative producers, who fear that their curves may be used, in spite of warnings to the contrary, for the setting of specifications, may also hug the lower limit of the band. In the curves shown in some advertising literature, the band might really be mostly below the line.

In the absence of sufficient data for other steels to allow band-plotting such as that for 6130 and 3130, most property charts available indicate what the person who prepares them assumes to be the *average* properties for the given steel. The accuracy of this assumption depends on the breadth of his experience and on his freedom from bias.

Property charts are indispensable in showing the general behavior of a class of steel, but line charts should be considered as *only a first approximation*. In general the property charts for individual steels shown herein will be taken from such sources as Bethlehem Alloy

Steels,<sup>23</sup> Nickel Alloy Steels,<sup>24</sup> and similar publications, as more representative of steels now on the market than the older S.A.E. curves. Both the Society of Automotive Engineers and the American Society for Metals are taking steps to secure data more completely representative of the steels currently supplied, and to pin down more closely the effect of variations in composition and grain size.

Data on C steels evaluated by probability methods are lacking save in rare cases. Sisco<sup>5</sup> has approached the problem by assembling single curves from various sources into bands, and in some cases picking therefrom the "most probable" values. In Fig. 4 his data are shown for water-quenched 1020 steel in  $\frac{1}{2}$ -in. to  $1\frac{1}{2}$ -in. diam., drawn as shown. (Only one set of data was plotted for yield strength.) Since Sisco's data were assembled, Dawe<sup>25</sup> has presented other data, which have been plotted in Fig. 4 as circles for comparison with Sisco's bands. It will be seen that while there is agreement as to general order of magnitude, the agreement is not exact.

Sisco<sup>5</sup> has given plots showing the bands for water-quenched S.A.E. 1035 and 1045 steels. Since the upper C range of 1035 and the lower of 1045 are both at 0.40% C, if one plots the extremes of those bands he should obtain representative data for 0.40% C. Sisco has also made a separate evaluation, from data for steels of 0.37-0.43% C, to show what he considers "best values." "Bethlehem Alloy Steels"<sup>23</sup> presents curves for 1040. These data are collected in Fig. 4.

The value of curves of this sort is in showing trends rather than

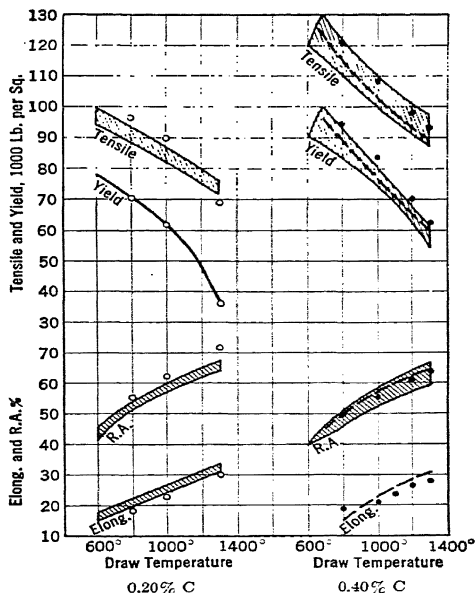


FIG. 4.—Properties of 0.20 and 0.40% C steels water quenched, small sections.

Bands, Sisco.<sup>5</sup>  
Circles, Dawe.<sup>25</sup>  
Dashed line, Sisco.<sup>5</sup> most probable.  
Points, Bethlehem.<sup>23</sup>



exact values. Even if one procures a steel from a certain maker, heat treats it according to the schedule used by that maker in preparing a set of property curves, and then reads his values from those curves instead of testing the actual steel, he has little assurance that the

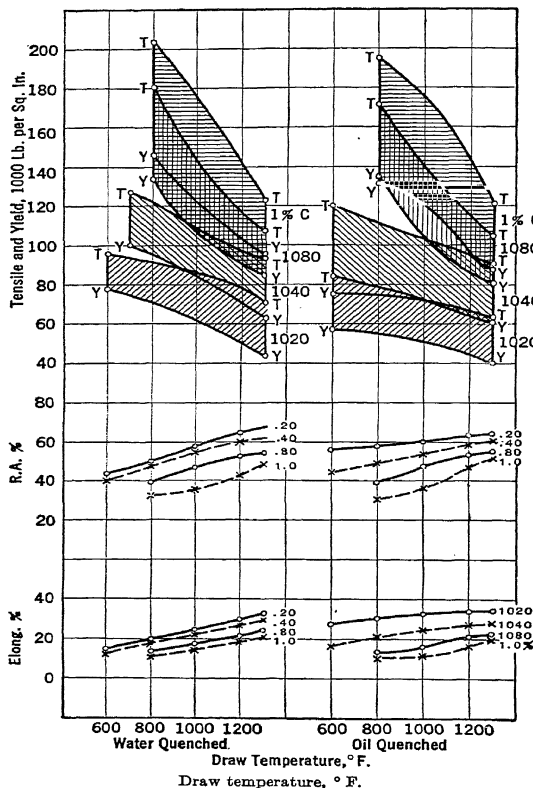


FIG. 5.—Average properties of 1-in. diam. bars of varying C content, water or oil quenched and drawn as shown.

values taken represent anything more than a general order of magnitude. It will often be necessary, in later chapters, to compare steels or treatments by property plots when data are lacking to establish the bands, but the reader should always remember that the curve is only a general indication and not a matter of exactness. We shall not ordi-

narily enter the hardness values on such curves since through the tensile values these can be found from the conversion curves of Figs. 7 and 8, Chap. 1, Vol. I. Too many lines upon these charts make them hard to read.

The trends of properties for water- or oil-quenched and drawn C steels for a wider range of C content are indicated in Fig. 5.

The variations in static ductility with C content noted in Fig. 4 would be exaggerated were similar data shown for impact. It will

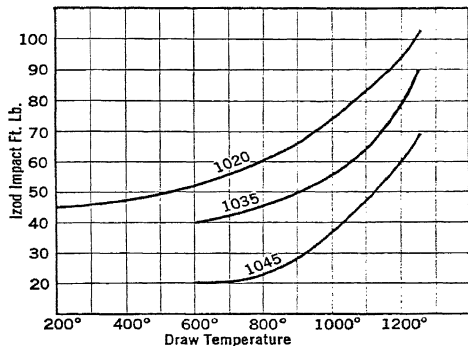


FIG. 6.—Impact resistance of carbon steels.

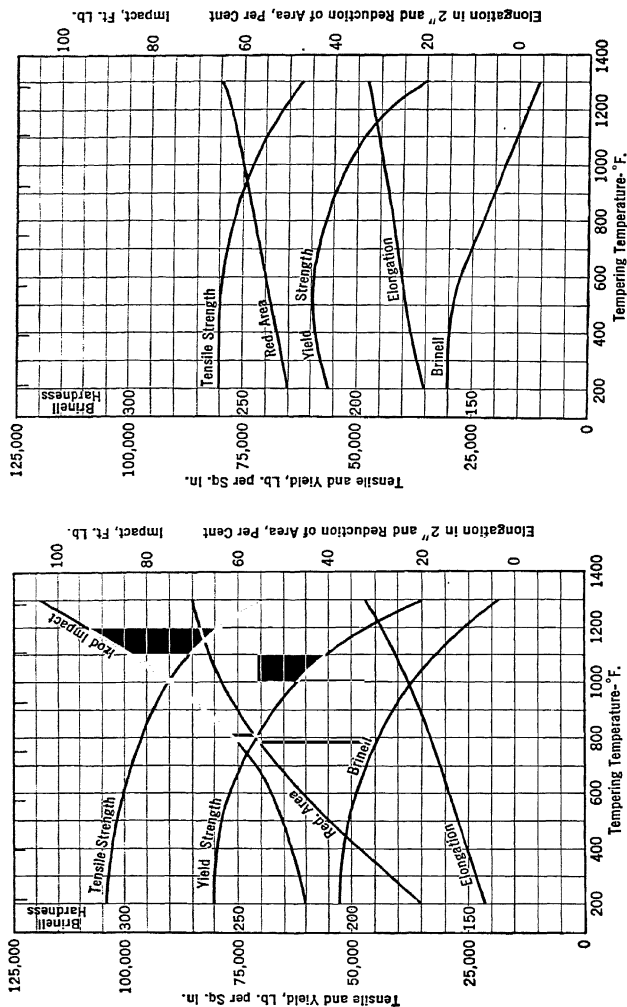
Water quenched as follows: 1020–1600°, 1035–1550°, 1045–1500°. Drawn as shown.  
Small sections.

also be remembered that impact is materially affected by grain size. The trend of behavior in impact is indicated by Fig. 6, in which the curves, collected from those of Figs. 7–9, are presumably comparable since they were all taken from one source.<sup>24</sup>

**Oil vs. Water Quenching.** The effect of water vs. oil quenching of C steels in small sizes in terms of average curves rather than as bands is shown in Figs. 7–9, for small sections,  $\frac{1}{2}$ - to  $1\frac{1}{2}$ -in. diam., as given in Nickel Alloy Steels.<sup>24</sup> Figures 10 and 11, for high-C steels, are from Bethlehem Alloy Steels.<sup>23</sup>

#### BIBLIOGRAPHY

1. American Foundrymen's Association: Report of Steel Division Committee on Test Coupons, *Trans. Am. Foundrymen's Assoc.*, v. 45, 1937, pp. 97–109.
2. SIMS, C. E., and F. B. DAHLE: Effect of Aluminum on the Properties of Medium Carbon Cast Steel, *Am. Foundrymen's Assoc.*, Preprint 38–22, May, 1938, 40 pages.
3. American Foundrymen's Association and the American Society for Testing Materials: "Symposium on Steel Castings." Phila., The Societies, 1932, 254 pages.



Water quenched from 1575-1625°.

Oil quenched from 1600-1650°.

Fig. 7.—Properties of S.A.E. 1020 steel, 0.15-0.25 C, 0.30-0.60% Mn. Small sections  $\frac{1}{2}$ - to  $1\frac{1}{2}$ -in. diam., quenched and drawn as shown. (Nickel Alloy Steels.<sup>24</sup>)

See general comment on property charts, p. 40.

# PROPERTY CHARTS

45

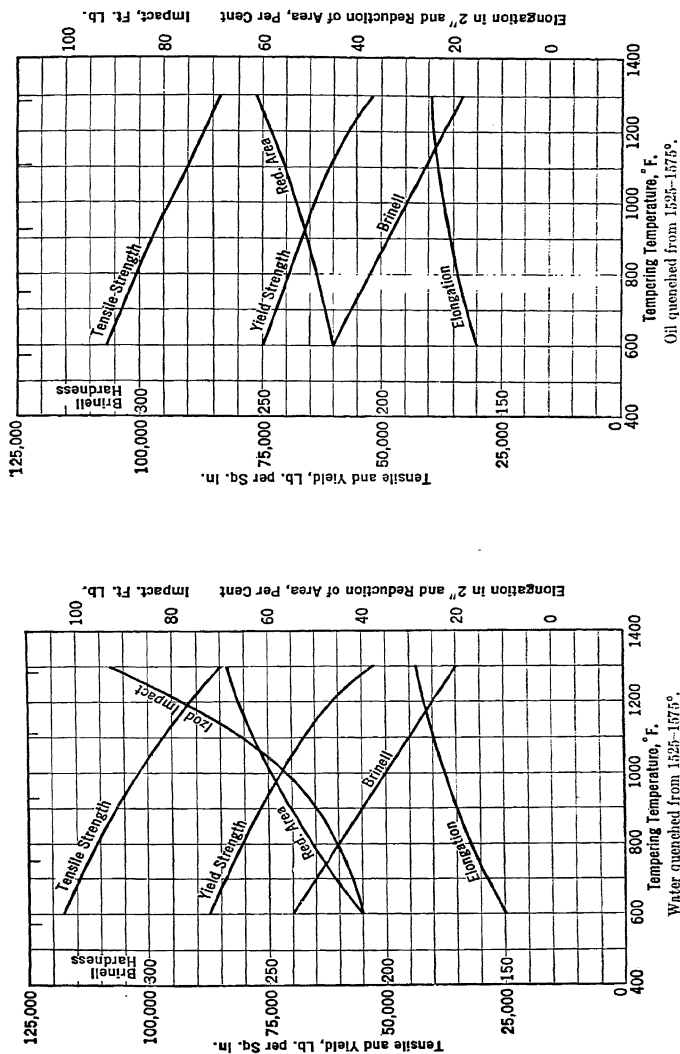


Fig. 8.—Properties of S.A.P. 1085 steel, 0.30-0.40 C, 0.50-0.80% Mn. Small sections, ½- to 1½-in. diam., quenched and drawn as shown: (Nickel Alloy Steels.<sup>24</sup>)

See general comment on property charts, p. 40.

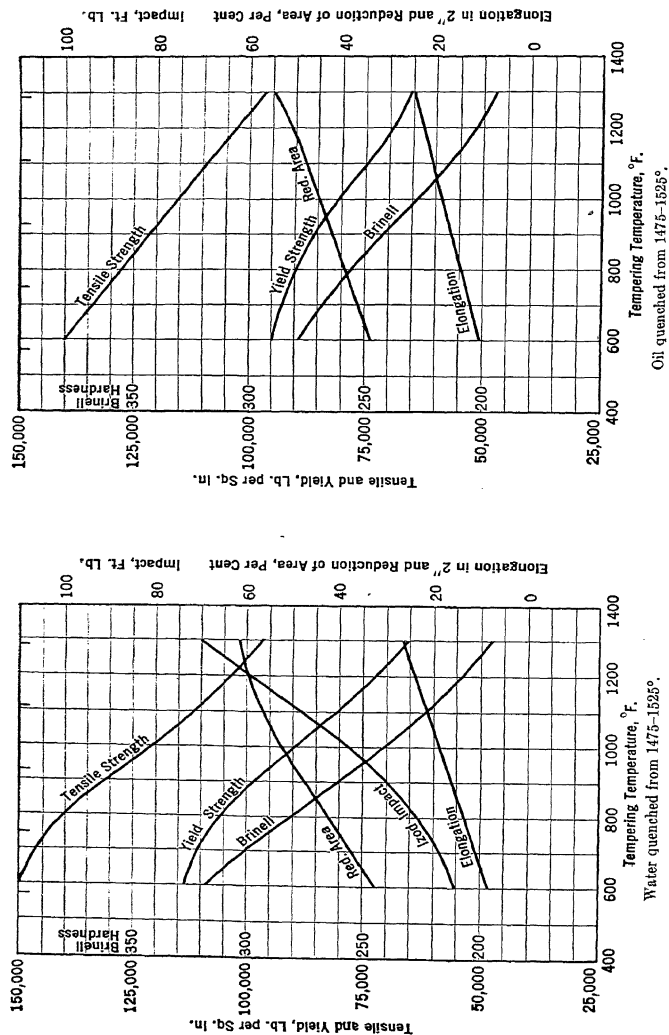


Fig. 9.—Properties of S.A.E. 1045 steel, 0.40-0.50 C, 0.50-0.80% Mn. Small sections,  $\frac{1}{2}$ - to  $\frac{1}{4}$ -in. diam., quenched and drawn as shown. (Nickel Alloy Steels.<sup>24</sup>)

See general comment on property charts, p. 40.





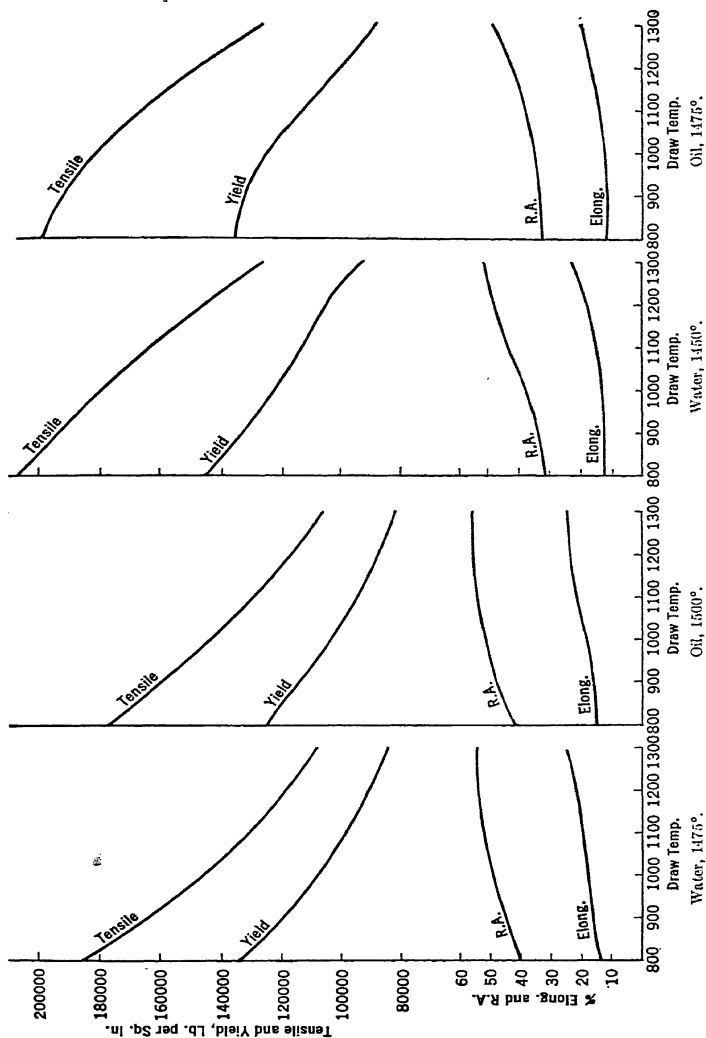


Fig. 11.—Properties of 1% C steel.

0.90-1.05 C, 0.25-0.50 Mn, 0.15-0.30% Si.

(Bethlehem Alloy Steels<sup>24</sup>)

See general comment on property charts, p. 40.

Fig. 10.—Properties of 0.80% C steel.

0.75-0.85 C, 0.30-0.50 Mn, 0.15-0.25% Si.

Steels first normalized from 1650°F, in 1-in. round

See general comment on property charts, p. 40.



4. Institution of Automobile Engineers and the Society of Motor Manufacturers and Traders: Report of the Steel Research Committee. Sheffield, October, 1920.
5. SISCO, F. T.: "The Alloys of Iron and Carbon." Vol. II: Properties. New York, McGraw-Hill Book Co., Inc., 1937, 777 pages.
6. Joint Committee on Effect of Phosphorus and Sulphur in Steel. Report, *Proc. Am. Soc. Test. Mat.* v. 22, part I, 1922, pp. 94-126.
7. SMITH, R. H.: Some Physical Properties of Low Carbon Steel, *Trans. Am. Soc. Steel Treat.*, v. 7, 1925, pp. 569-580.
8. SMITH, R. H.: High Tensile Strengths with Low-Carbon Steels, *Proc. Am. Soc. Test. Mat.*, v. 24, part 2, 1924, pp. 618-625.
9. EPSTEIN, S.: "The Alloys of Iron and Carbon." Vol. I: Constitution. New York, McGraw-Hill Book Co., Inc., 1936, 476 pages.
10. HERTY, C. H., JR., and B. N. DANILOFF: The Effect of Deoxidation on the Aging of Mild Steels, *Cooperative Bull.* 66, Min. Met. Advisory Boards to Carnegie Inst. Tech. and Bur. Mines, Pittsburgh, 1934, 42 pages.
11. GRIFFIS, R. O., R. L. KENYON and R. S. BURNS: The Aging of Mild Steel Sheets, *Yearbook Am. Iron Steel Inst.*, v. 23, 1933, pp. 142-160, discussion, pp. 160-165.
12. LORIG, C. H., and C. E. WILLIAMS: Physical and Mechanical Properties of Some Well-Known Cast Steels, A.F.A.-A.S.T.M. "Symposium on Cast Steels," 1932, pp. 76-158.
13. SWANGER, W. H., and G. F. WOHLGEMUTH: Failure of Heat-treated Steel Wire in Cables of the Mt. Hope, R. I., Suspension Bridge, *Proc. Am. Soc. Test. Mat.*, v. 36, part 2, 1936, pp. 21-84.
14. WISEHART, H. B., and W. D. BOONE: Properties of Rail Steel at Arctic Temperatures, *Metal Progress*, v. 29, January, 1936, pp. 61-62.
15. CRAMER, R. E.: Production of Flakes in Steel by Heating in Hydrogen, *Trans. Am. Soc. Metals*, v. 25, 1937, pp. 923-934.
16. GOODAIRE, E. W.: Heat Treated Rails for Special Trackwork, *Metal Progress*, v. 29, March, 1936, pp. 59-64, 92.
17. KENNEY, E. F.: Heat Treatment of Rails. U.S. Patent 1,846,644, February 23, 1932.
18. SANDBERG, C. P., O. F. A. SANDBERG, and N. P. P. SANDBERG: Effect of Controlled Cooling and Temperature Equalization on Internal Fissures in Rails, *Metals & Alloys*, v. 3, 1932, pp. 89-92.
19. New Furnace Treats Rails for Higher Ductility and Impact Resistance, *Steel*, v. 100, Mar. 15, 1937, p. 74.
20. DICK, M. H.: How the Welding Was Done, *Ry. Eng. & Maintenance*, v. 32, January, 1936, pp. 17-20.
21. BRONSON, C. B.: Rail End Building Up and Heat Treatment by the Arc Method, *Weld. J.*, (N.Y.), v. 16, January, 1937, pp. 12-13.
22. MOORE, H. F.: Third Progress Report on the Investigation of Fissures in Railroad Rails, *Bull. Am. Ry. Eng. Assoc.*, v. 39, March, 1937, pp. 57-85.
23. Bethlehem Steel Company: "Bethlehem Alloy and Special Steels." Bethlehem, Pa., The Author, 1935, 375 pages.
24. FRENCH, H. J., and J. W. SANDS (editors): "Nickel Alloy Steels." New York, International Nickel Co., Inc., 1934, 7 sections.
25. DAWE, C. N.: "The Steel (Physical Property) Atlas." Cleveland, American Society for Metals, 1936, 87 pages.

## CHAPTER 2

### PHOSPHORUS, SULPHUR, LEAD, AND FREE-MACHINING STEELS

**Phosphorus.** The elements P, S, Mn, and, in killed steels, Si, are always present in C steels. Although the alloying effects of Mn and Si are fully recognized, P and S have been generally considered as "nuisance" elements and held to low figures. In the S.A.E. steels other than the free-cutting types, P is held to 0.03–0.045% max., S to 0.03–0.055% max. Yet both P and S are intentionally introduced into certain grades of steel, especially those to be used without quenching and tempering.

**Properties Conferred by Phosphorus.** If equal amounts of C and P are added to Fe, or if to a base steel of say 0.10% C and 0.02% P, either more C or more P is added, it is found that the addition of P is the more potent, that for a given tensile strength the high P steel has a higher yield strength, and, up to quite high P, as good or better static ductility than the equivalent C steel, i.e., P is a strong ferrite-strengthener. It does not enter the carbide. As long as the sum of C+P does not rise much above 0.30% the impact resistance at room temperature is not badly reduced. At about that sum the impact falls sharply, though static ductility, and even ability to withstand repeated bending, are maintained to much higher P contents. The impact may be improved by making the grain size of the steel small, by suitable alloying and "deoxidizing" additions.

The permissible amount of P is strictly limited by the C content; so P as an alloying element is utilized only in low C steels, normally of 0.10–0.15% C max. Hence quenching and tempering do not enter the picture in the P-alloyed steels.

**Embrittlement by Phosphorus.** These effects are shown in Fig. 1 from Lorig and Krause.<sup>1</sup> Below room temperature, the embrittling effect of C and P is accentuated. Hence, 0.25% C+P seems to be the upper limit for safety, though by the use of Al for grain-size control, addition of Cr or other helpful elements, and holding Si (which accentuates brittleness) to a low limit, it might be feasible to exceed this, were it not for the danger of segregation.

**Segregation.** A P-containing steel, on slowly cooling in the ingot, is prone to segregation of P, and segregated P-bearing ferrite is not readily homogenized; so such steels when rolled are likely to show a banded structure that is not readily broken up, and to have patches higher in P than the average composition of the steel as a whole.

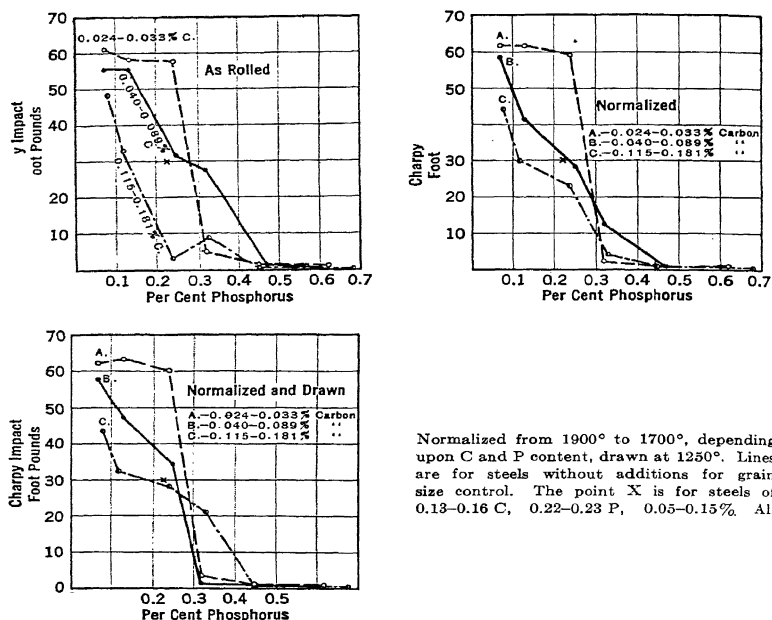


Fig. 1.—Effect of C, P, and of heat treatment upon impact. (Lorig and Krause.<sup>1</sup>)

Hence the P is kept down so that any P-rich areas will not be too high in P for impact resistance.

Some specifications for Bessemer steel allow 0.10% P max. In a footnote to A.S.T.M. specification A139 : 36 dealing with electric fusion welded steel pipe, it is stated that commercial welding practice avoids more than 0.35% C in open-hearth steel, or more than 0.15% C in Bessemer steel. This limit for avoidance of brittleness in welds agrees with the 0.25% P+C limit mentioned above for avoidance of brittleness in impact.

The embrittling effect of P is thought to add to the temper brittleness often noted in Cr, Mn, and Ni-Cr steels; so this is another reason

for holding it low in steels with those alloying elements, when they are to be quenched and tempered.

**Yield Strength.** Notwithstanding these limitations, P, intelligently used, is far from being solely a nuisance element. It has marked value in producing a higher yield strength than would be obtainable in a C steel of this class. Figure 2<sup>2</sup> shows the relation between P content and tensile and yield strengths as well as a comparison of the effect of P and C.

**Corrosion Resistance.** An even more valuable and specific effect of P is the improvement it confers in resistance to atmospheric corrosion, this effect being clearly shown even in the absence of other alloying elements, and very markedly in conjunction with Cu. These factors have led to the use of P as a definite alloying element in several of the newer "high yield strength, low-alloy" steels, which are discussed further in Chap. 11. For other details the reader is referred to recent summaries.<sup>1, 2, 3</sup>

**Special Effects.** Another little-realized specific effect of P is in raising high-temperature strength, as Speller<sup>4</sup> intimated and Cross and Krause<sup>5</sup> showed in detail.

In annealing or normalizing, it is necessary to remember that P sharply raises  $A_{c3}$ . In an 0.05% C steel, 0.15% P raises  $A_{c3}$  to about 1850°, from below 1700° without the P. In an 0.15% C steel, 0.15% P raises  $A_{c3}$  from about 1550° to nearly 1650°. Thus P raises  $A_{c3}$  quite as sharply as C lowers it.

The raising of  $A_{c3}$ , plus the strengthening effect of P at elevated temperatures, makes steel high in P useful as a base for vitreous enameling. Ordinary enameling stock is likely to sag during the firing of a high-temperature enamel, and especially when the firing temperature reaches  $A_{c3}$ . Krause<sup>6</sup> finds that 0.10 to 0.35% P raises  $A_{c3}$  sufficiently to avoid the distortion that occurs when a low-P steel passes through the critical, and hence prevents sagging.

**The Gamma Loop.** Although the amount of P in steels for quenching and tempering is too low to have much effect on  $A_{c3}$ , a similar

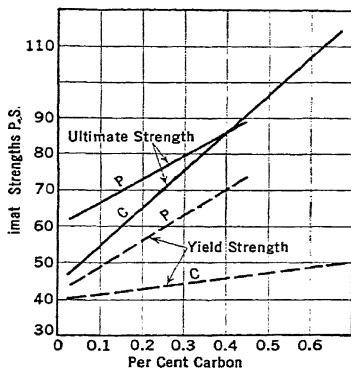


FIG. 2.—Effect of C and of P on strength of normalized Fe-C or Fe-P alloys. The Fe-P alloys contained about 0.10% C. (Epstein, Nead and Halley.<sup>2</sup>)

See general comment on property charts, p. 40, chap. 1.

effect does have to be considered with other elements, and we may here digress in order to comment on the general effect.

The effect of alloying elements upon the position of the  $Ac_3$  critical on heating must obviously be given careful attention in heat treating such steels. Unless the steel is heated high enough to make it fully austenitic, it is not completely refined; and, if it is heated unnecessarily far into the austenitic range, undesired grain coarsening may ensue. Whether the steel is to be normalized or quenched, the proper heating temperature will in each case bear a definite relation to  $Ac_3$ . It must be above  $Ac_3$ , though just how far above it depends on conditions.

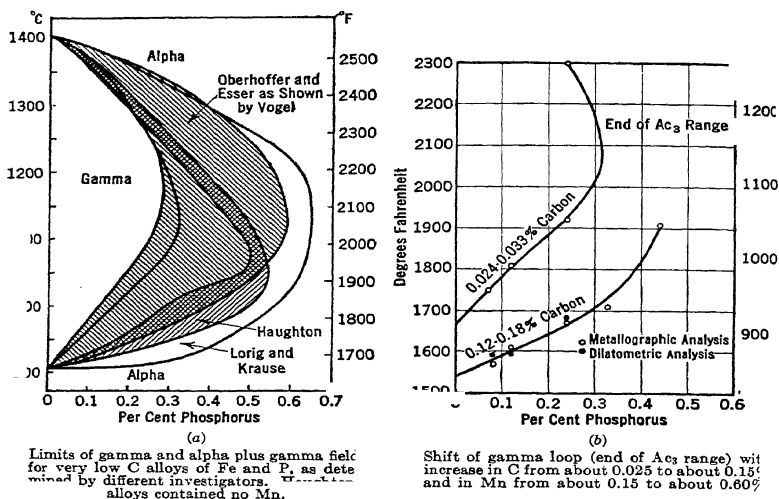


FIG. 3.—The gamma loop in the FeP system. (Lorig and Krause.)

Some elements, like Mn and Ni, lower  $Ac_3$ , others raise it and form a "gamma loop," i.e., a restricted area beyond which gamma iron (austenite) does not appear. The position of the loop is often materially altered with change in C content. Figure 3 shows the approximate location of the gamma loop for P steels.

**Bessemer Steel.** The low-C Bessemer steels with high P not only have a higher yield strength and are less "gummy" than analogous steels on machining, but when cold worked, as in the operations of removing and bending a chip, they work harden and embrittle more readily. Although the P content is not the only factor (since they also

contain more N than open-hearth steels), Bessemer steels, with their high P content, are recognized as having remarkable machinability and suitability for screw stock. The S.A.E. free-cutting series includes the following:

	%C	%Mn	%P	%S
1112.....	0.08-0.16	0.60-0.90	0.09-0.13	0.10-0.20
X1112.....	0.08-0.16	0.60-0.90	0.09-0.13	0.20-0.30

**Sulphur.** As was noted above, the free-cutting steels are high in S. Besides the high P, high S, 1112 steels, the S.A.E. lists others, with 0.045% P max. as follows:

	%C	%Mn	%S
1115.....	0.10-0.20	0.70-1.00	0.075-0.15
1120.....	0.15-0.25	0.60-0.90	0.075-0.15

Still others with higher Mn will be mentioned in Chap. 3.

**Free-cutting Steels.** That the lower S limit in these steels is 0.075% does not mean that perceptible influence on machining is not exerted by amounts below that limit. Peterson<sup>7</sup> comments that in 1045 steel for automotive crankshafts, raising the S to 0.055% (the upper limit of the usual S.A.E. specification for that steel) promotes better machining without detriment to physical properties.

The effect upon physical properties of S in low- and high-C steels or, more strictly, of S upon soft and hard steels, will be discussed under the Mn-alloyed, free-machining steels, which form another group, X1314-X1340, in the S.A.E. free-cutting steels and are so high in Mn as to be classed with the Mn steels, as the prefix 13 shows. The 1115 and 1120 grades are not high in Mn that is free to strengthen ferrite and enter the carbide, since quite a proportion of it is used in combining with the high S to form MnS inclusions so that these grades are scarcely any more responsive to heat treatment than the 1025 and 1020 steels, and generally are not intended to be heat treated.

**"Chip-breakers."** The MnS inclusions, both in the 1100 and the X1100 series, are introduced to give "stepping stones" for the advancing crack that travels ahead of the tool tip in machining, and for the cracking of the chip as it curls away from the tool. Although not listed in the S.A.E. series, steels with still higher S, up to around 0.50%, are made and used for special free-machining needs. Sometimes selenium, which forms selenides analogous to the sulphides, is employed, and instead of increasing the Mn to provide MnS (by combining with S,) Mo, or Zr, is utilized to form molybdenum or zirconium sulphides. The principle is the same, but detailed discussion here would lead us too far afield.

**Other Insoluble Additions.** In nonferrous alloys, the addition of small amounts of elements that are insoluble in the solid alloy (or that form compounds that are insoluble) and are dispersed throughout the alloy is widely used to increase machinability. Lead brass is a familiar example. Small insoluble additions including lead have recently been utilized in increasing the machinability of Al alloys. Little use has hitherto been made in steel of insoluble additions other than those that produce sulphides and other compounds of that group, though analogous benefits might be found. A benefit to machinability of high-Cr steel is obtained through the addition of Cu in amount so far beyond the solid solubility limit that globules of Cu are scattered through the steel.<sup>8</sup> In this particular case the addition of so much foreign material is not without its drawbacks in other ways.

Probably the most readily machinable ferrous material is regular malleable iron in which there is a fine dispersion of tiny particles of "temper carbon," while even the coarse graphite of ordinary cast iron greatly aids machinability. The effect of temper carbon is also shown in the easy machining properties of "graphitic" steel (see Chap. 4).

Very hard particles, such as clumps of  $Al_2O_3$  in "dirty" Al-treated steels are ruinous to tools. If the particles are of the degree of submicroscopic fineness that is postulated for the  $Al_2O_3$  particles in fine-grained steels (see Chap. 9, Vol. I) they are not notably harmful to tools, but they are not effective in causing the chips to break up and do not aid machinability.

**Lead.** A new addition<sup>9</sup> to the family of free-machining steels is produced by the use of lead. Lead is practically insoluble in liquid or solid steel, but by suitable methods of addition, a small amount can be suspended in steel. The particles of Pb are in such a fine state of subdivision that they are not visible under the microscope on an unetched section, and special etching methods are required to show their location. Doubtless because of this fine state of subdivision, the Pb does not appreciably affect the mechanical properties of the steel, as the following data for cold-drawn screw stock indicate:

C	Mn	Si	P	S	Pb	Tensile	Yield	Elong. %	R. A. %	Charpy
0.11	0.63	0.01	0.017	0.193	none	75,000	73,500	17	55.5	23
0.10	0.55	0.01	0.019	0.204	0.12	73,500	72,000	18.5	56.5	20
0.14	0.68	0.25	0.019	0.197	none	78,500	76,750	16.5	52.5	19.5
0.11	0.68	0.24	0.020	0.207	0.09	77,000	75,250	17.5	53	22

In each of these the Pb addition produced about 20% increase in machinability.

The Pb appears to act as an internal lubricant in machining, and probably adds to the chip-breaking ability. Material increases in speed of machining are reported for steels containing 0.10–0.25% or more Pb. A high-S steel that is already free-machining has its machinability increased by the Pb addition. Steel low in P and S may be brought to a level of machinability much the same as that of the standard free-cutting steels, by use of Pb.

From the point of view of heat treatment, the Pb addition is of particular interest, since the properties are so little affected by this addition. For example, hot-rolled S.A.E. 1020 shows the following properties:

	Tensile	Yield	Elong. %	R. A. %	Charpy	Brinell
Hot-rolled S.A.E. 1120, no Pb....	58,000	39,500	31	56	36.5	106
Hot-rolled S.A.E. 1120, same heat, 0.25% Pb added.....	58,500	39,500	28.5	51	35	106

In a harder steel, of 0.90 C, 0.15 Si, 0.70% Mn, annealed at 1450°, the following properties were obtained:

	TENSILE	ELONG. %	R. A. %	BRINELL
No Pb.....	141,500	10.5	20	269
0.20% Pb added....	139,500	12.5	24.5	277

The second steel showed more than 20% improvement in machinability over the first.

In a still harder Ni-Cr-Mo steel, oil quenched and tempered to 341 Brinell, bars containing 0.16% Pb machined 25% more readily than those of the same hardness without Pb. From the point of view of testing it is of interest, that the relative machinability of the Pb-free and the Pb-containing steels was determined in the laboratory by simple drilling and sawing tests, and that the increase in machinability indicated by these tests was later established in screw-machine operation.

The effects of Pb on the mechanical properties of steel are further illustrated in the data tabulated below. In the first group, data are presented for low-C alloy steels tested in the normalized condition. The S.A.E. number gives the type composition and the second column



gives the Pb content. As is evident from the data presented the Pb had practically no effect on the tensile strength, ductility, hardness and impact resistance.

Machinability tests on these steels in which comparisons were made between the Pb-free or low-Pb steels and those containing appreciable amounts of Pb, showed improvements in machinability ranging from 19 to 42%, the average being about 25%.

MECHANICAL PROPERTIES OF LOW CARBON ALLOY STEELS AS NORMALIZED \*

S.A.E. Type	Pb %	Normalizing Temp. ° F.	Tensile Strength	Yield Strength	Elong. %	R. A. %	Brinell Hardness	Charpy Impact ft. lbs.
2315	0.01	1500	86,500	58,000	31.0	63.0	163	44.0
2315	0.17	1500	88,000	62,500	31.5	61.0	163	39.0
3115	0.00	1600	79,500	57,000	33.0	70.0	146	55.5
3115	0.14	1600	79,000	60,500	34.0	67.5	149	49.0
4320	0.03	1600	115,000	80,000	21.0	57.0	207	35.0
4320	0.11	1600	115,000	77,500	21.0	56.5	207	33.0
4615	0.05	1600	86,500	52,000	30.5	63.5	163	47.0
4615	0.18	1600	86,000	52,500	28.0	62.5	156	43.0
5120	0.02	1700	76,000	56,000	35.0	68.0	131	48.0
5120	0.12	1700	75,500	54,000	35.0	66.0	137	51.0
6115	0.00	1700	75,000	59,500	35.0	73.0	131	58.5
6115	0.16	1700	75,000	58,000	35.5	70.5	126	53.0

\* Private communication.

Data for oil quenched and drawn higher C alloy steels are presented in the Table on page 57.

The correlation between the hardnesses and the strengths is obvious in the above tabulation.

Machinability tests on these steels heat treated to the same hardnesses showed improvements of the Pb-containing steels as compared with the Pb-free or low-Pb steels of 4 to 40%; the average improvement in machinability being 25%.

Although the Pb addition is too new for its full range of application to be definitely demonstrated as yet, it is possible that the general level of machinability for heat-treated C and alloy steels may be

## MECHANICAL PROPERTIES OF HIGH CARBON ALLOY STEELS AS OIL QUENCHED AND DRAWN \*

S.A.E. Type	Pb %	Quenching Temp. ° F.	Drawing Temp. ° F.	Tensile Strength	Yield Strength	Elong. %	R. A. %	Brinell Hardness	Charpy Impact ft. lbs.
2345	0.02	1475	1150	125,500	85,500	20.0	42.5	223	23.5
2345	0.20	1475	1150	124,000	94,000	20.0	45.5	229	23.0
3145	0.03	1475	1200	156,000	143,000	19.5	58.0	293	29.5
3145	0.18	1475	1200	151,000	140,000	19.0	58.5	285	30.0
4130	0.00	1600	1125	133,000	117,500	20.0	65.0	255	45.0
4130	0.12	1600	1125	119,000	99,000	20.5	61.6	217	44.0
4340	0.03	1525	1200	151,000	134,000	18.0	56.5	272	37.5
4340	0.17	1525	1200	142,000	115,500	18.0	53.0	250	33.5
4640	0.05	1525	1125	147,500	133,000	20.0	61.0	265	35.0
4640	0.18	1525	1125	140,000	120,000	20.0	58.0	255	35.0
6150	0.03	1550	1200	164,500	154,500	18.0	55.5	311	24.5
6150	0.21	1550	1200	163,500	148,000	16.5	53.0	321	23.5
T1330	0.03	1525†	1125	122,500	113,000	19.0	58.0	226	21.5
T1330	0.15	1525†	1125	121,000	113,000	19.5	55.0	226	20.0
5150	0.04	1550	1200	152,500	131,000	19.0	57.5	282	30.0
5150	0.17	1550	1200	150,000	127,000	18.0	57.0	285	28.0

\* Private Communication.

† Water Quenched.

raised by the lead addition; that is, it may be feasible to machine steels of greater tensile strengths and hardnesses than those at which production machining has hitherto been considered practical.

**Cold Working.** The free-machining steels, normally employed in rather small sections and especially for parts made on automatic screw machines, are generally work-hardened by cold drawing. This work-hardening increases the reduction of area in the tensile test.

The relationship of reduction of area in the tensile test to machinability is thought to lie in the fact that, in a steel which tends to deform locally by necking, rather than generally by uniform stretching the distortion and consequent work hardening on the surface of the chip, and on that from which the chip is being removed, are localized. Compare Fig. 13, Chap. 2, Vol. I. Hence the crack that advances

ahead of the tool point progresses more easily without using energy in distortion of underlying layers.

The cold drawing of bars for screw machine use reduces the general elongation and makes the steel neck down locally. As an analogy, it may be noted that in rather hard heat-treated steels containing Mo, the reduction of area is greater than in steels of similar hardness without Mo. It is generally believed that the Mo steels are machinable in production at a hardness at which steels without Mo would be classed as unmachinable.

**Structure for Machinability.** A very ductile "gummy" metal has poor machinability; so the very soft, low-C ferritic steels are avoided for rapid machining. If the ferrite is sufficiently interspersed with harder, more brittle, particles, like  $\text{Fe}_3\text{C}$ , the ferrite patches tend to crack more readily, the crack advancing through the ferrite from one  $\text{Fe}_3\text{C}$  patch to another. Compare Armour.<sup>10</sup> Hence increasing the C content from 0.10 to 0.20% improves machinability in spite of the fact that the latter steel is the harder and stronger. The most favorable arrangement of ferrite and cementite to give short ferrite paths for the crack is in lamellar pearlite; so a heat treatment that will produce lamellar pearlite is advantageous for machinability. Of course, if the C content is raised very high, as in eutectoid and hypereutectoid steels, mere strength and hardness may make it too difficult to force the tool through, and some degree of spheroidizing anneal may be required to offset this. The influence of the size and distribution of the spheroids has been discussed by Roda.<sup>11</sup>

Naturally, the alloyed steels are more difficult to machine than plain C steels, because the alloyed ferrite is tougher and the complex carbides are generally harder. The requirements for ready machinability are weakness and brittleness at the temperature the steel reaches right at the point of chip separation. As the temper colors on a steel chip removed in rapid machining show, this temperature is quite elevated; so the room temperature properties do not necessarily accurately reflect behavior in machining.

**Effect of Alloys on Machinability.** Alloying elements and strengthening heat treatments shift the properties of steel in the direction desired from the point of view of load-carrying ability, but usually in the wrong direction for machinability; so a compromise has to be effected. With modern grinding methods and with the availability of the tungsten carbide type of tool, the finishing of very hard heat-treated steels has become commercial and much of the former necessity for compromise removed. But it is always desirable to leave as little stock as possible to be removed in the very hard state; so rough

machining must often be done prior to final heat treatment, and the compromise enters at this stage.

According to the "A.S.M. Handbook," if a free-cutting 1112 steel of 0.15 S, 0.10% P is taken as 100 in machinability, the higher S, X1112 has a machinability of 140; the low-P, high-Mn X1314 has 90-95; the higher C X1335 and X1350, 70-75; whereas the C forging grade steels and the usual carburizing steels run 50-65, and the higher alloy groups run 30-50, all in the cold-drawn condition. The difficulty of testing machinability has been mentioned in Chap. 1, Vol. I.

**Nitrogen.** Nitrogen in very small but appreciable amounts is present in Bessemer steel, and it is sometimes intentionally added to open-hearth steel for free-machining uses. It has a hardening effect somewhat analogous to that of P.

**Heat Treatment.** The low-C, free-machining steels are not commercially hardened by quenching, but, as was discussed in Chap. 1, the presence of MnS inclusions is not fatal in soft steels for most uses; so the 1120 steels, which respond slightly to quenching, are sometimes utilized where the mechanical properties of quenched and tempered 0.20% C steel are sufficient, and good machinability is called for.

The 1100 series is not usually chosen for carburizing stock. When response to quenching or ability to be readily carburized is sought in free-machining steels, the Mn-alloyed X1300 series is commonly employed.

**Wide Use of Free-machining Steels.** Although much more has been written about the production and use of other grades of steel than about the free-machining grades, the latter are very widely used. One steel producer made a study of the amount of high-grade, wrought-steel bar stock used and concluded that "soft steel" amounted to 20%; the S.A.E. plain C 1000 series of steels to 13%; free-cutting steels to 20% (of which about 15% is in the 1100 series, mostly 1112, and about 5% in the X1300 or Mn free-cutting series). That the free-cutting steels are used in greater amount than all the carburizing, forging, and spring steels of the plain C grade bar stock put together emphasizes their usefulness.

## BIBLIOGRAPHY

1. LORIG, C. H., and D. E. KRAUSE: Phosphorus as an Alloying Element in Low Carbon Low Alloy Steels, *Metals & Alloys*, v. 7, 1937, pp. 9-13, 51-56, 69-73. See also LORIG, C. H.: High Phosphorus Steels, *Metal Progress*, v. 33, July, 1938, pp. 60-61.
2. EPSTEIN, S., J. H. NEAD, and J. W. HALLEY: Choosing a Composition for Low-Alloy High-Strength Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 120,

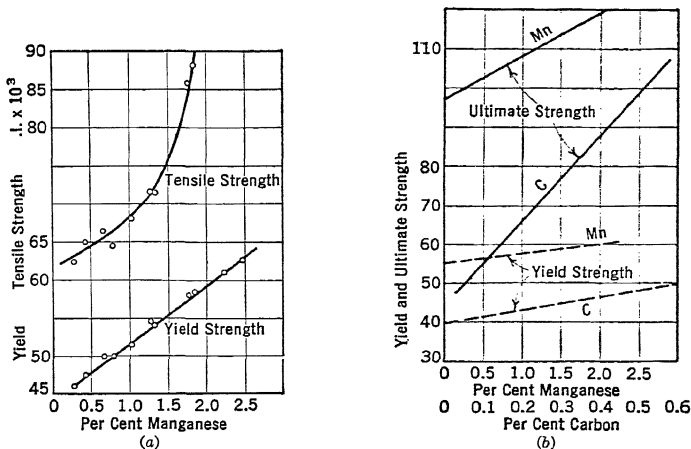
## PHOSPHORUS. SULPHUR. LEAD, FREE-MACHINING STEELS

- 1936, pp. 309-345. Also: *Metals Technology*, v. 3, 1936, Tech. Publ. 697, 31 pages.
3. GILLETT, H. W.: Phosphorus as an Alloying Element in Steel, *Metals & Alloys*, v. 6, 1935, pp. 280-283, 307-310 (Corr. Abst.).
4. SPELLER, F. N.: Discussion. *Proc. Am. Soc. Test. Mat.* v. 24, part 2, 1924, pp. 162-163.
5. CROSS, H. C., and D. E. KRAUSE: Phosphorus as an Alloying Element in Steels for Use at Elevated Temperatures, *Metals & Alloys*, v. 8, 1937, pp. 53-58.
6. KRAUSE, D. E.: Vitreous Enameled Product, U.S. Patent 2,109,271, February 22, 1938.
7. PETERSON, O. N.: New Concepts of Metallurgical Control in Automotive Steels, *Steel*, v. 100, June 14, 1937, pp. 36-39, 72.
8. LIPPERT, T. W.: 10 Cu + 18 Cr + 72 Fe, etc. = Digby's Cypritic Alloy, An Entirely New Species of the Genus Stainless Steel, *Iron Age*, v. 140, October 14, 1937, pp. 54-63.
9. Inland Steel Co.: New Free-cutting Steel Introduced. *Steel*, v. 102, May 16, 1938, p. 29.
10. ARMOUR, J. D.: Improving the Free-Cutting Steels, *Metal Progress*, v. 33, January, 1938, pp. 60-64. Also: Machinability of Steel, *Heat Treat. Forg.*, v. 24, 1938, pp. 130-134, 142.
11. RODA, D. F.: Relation of Size of Spheroids in Tool Steel to Its Machinability and to Holding Edge of Cutter. *Am. Soc. Metals*, Preprint 10, 1937, 9 pages.

## CHAPTER 3

### MANGANESE CONSTRUCTIONAL STEELS

**Alloying for Heat Treatment.** The principles of heat treatment are identical whether C or alloy steels are dealt with. But each alloying element has specific effects of which advantage may be taken. Combining two or more alloying elements can bring further advantage. Each alloy steel deviates somewhat from any other in its response to



Normalized steels. The Mn steels of this plot are very low in C.

Normalized steels. The Mn steel contains 0.55% C.

FIG. 1.—Strengthening effects of Mn and C. (Epstein, Nead and Halley.<sup>1</sup>)

See general comment on property curves, p. 40, chap. 1.

heat treatment, so it is necessary to consider the effects, first of each important element separately, then in combination. This will be done in this and subsequent chapters, starting with the effects of increased amounts of the elements that are used in plain C steel.

When a harder and stronger steel is needed, and for some reason it is inadvisable to increase the C to produce these properties, alloying elements are resorted to. Of two comparable steels, on the same level as to strength, one a plain C steel, the other an alloy steel, the latter

is always of the lower C content. Carbon is subtracted when alloy is added. If this is not done the optimum combination of strength and toughness is not reached.

The reasons for using a lower C plus alloy instead of a higher C may be many and varied. A common reason is that the higher C steel may have too low a yield ratio.

**Manganese.** Manganese is probably the most widely used alloying element, because of its cheapness and because it is present in all steels in small amounts (as a deoxidizer and to combine with S), so that utilization of scrap with increased Mn introduces no complications.

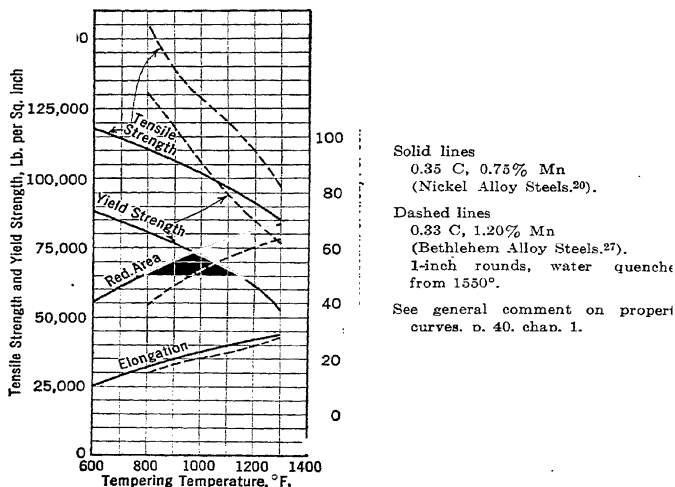


FIG. 2.—Effect of increase in Mn at the same C content.

For each grade of C steel there is an analogous grade with a little less C and a little more Mn.

**Medium Manganese Steels.\*** Manganese improves the yield ratio as Fig. 1<sup>1</sup> shows for normalized steels and Fig. 2 for those water quenched and tempered. In Fig. 2 it will be noted that the addition of Mn has brought the yield strength up around the tensile strength of the plain C steel, without excessive loss in ductility. That this is more effectively accomplished than by merely raising the C may

\* Only pearlitic, i.e., "medium" Mn steels are discussed in this chapter. Austenitic Mn steels are discussed in Chap. 20.

seen from Fig. 3, in which data for oil-quenched 0.33 C, 1.20% Mn steel are compared with data of Fig. 9, Chap. 1, for oil-quenched S.A.E. 1045. Similar differences are shown in a comparison of water-quenched specimens of the same steels. For a given tensile strength the ductility remains practically the same while the yield strength is materially raised by the Mn.

To show the effect of increasing the Mn in oil-quenched and tempered steel of a given C content, Fig. 4, the data for which are collected from several sources, is included. Figure 5 shows data for water-quenched and tempered steels of 0.37 C, 1.80% Mn, and 0.40 C, 1.15% Mn, assembled from separate curves given in earlier editions.

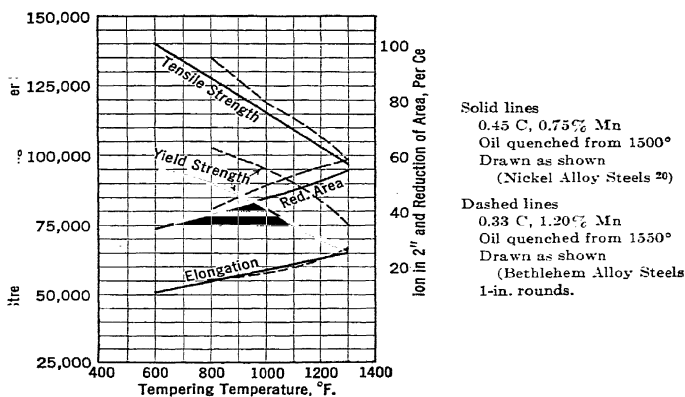


FIG. 3.—Effect of raising Mn and lowering C.

See general comment on property curves, p. 40, chap. 1.

**Depth-hardening.** Perhaps the worst failing of C steels to be quenched and tempered is their lack of depth hardening. Manganese adds strong depth-hardening properties. By raising the Mn of a 0.40% C steel by 1% the critical cooling rate is dropped from 1000°/sec. to 90°/sec. This action will be discussed in Chap. 10, Figs. 5, 6, and 9 of which show how effectively Mn increases the depth of hardening.

**Ar<sub>1</sub> and Ar''.** The effect of Mn in increasing hardenability is shown by the lowering of Ar'' (compare Chap. 6, Vol. I), as well as by the lowering of the critical cooling rate. Scott and Hoop<sup>2</sup> express the mutual effect of Mn and C upon austenite to make it sluggish, and to depress Ar'', plotted on the basis of equivalent C or Mn content, in Fig. 6. This effect would be complicated by variations in grain size.



## MANGANESE CONSTRUCTIONAL STEELS

The same thing is brought out in another way by the evaluation of the retarding effect of Mn on the  $Ar_1$  transformation of austenite at constant temperature.

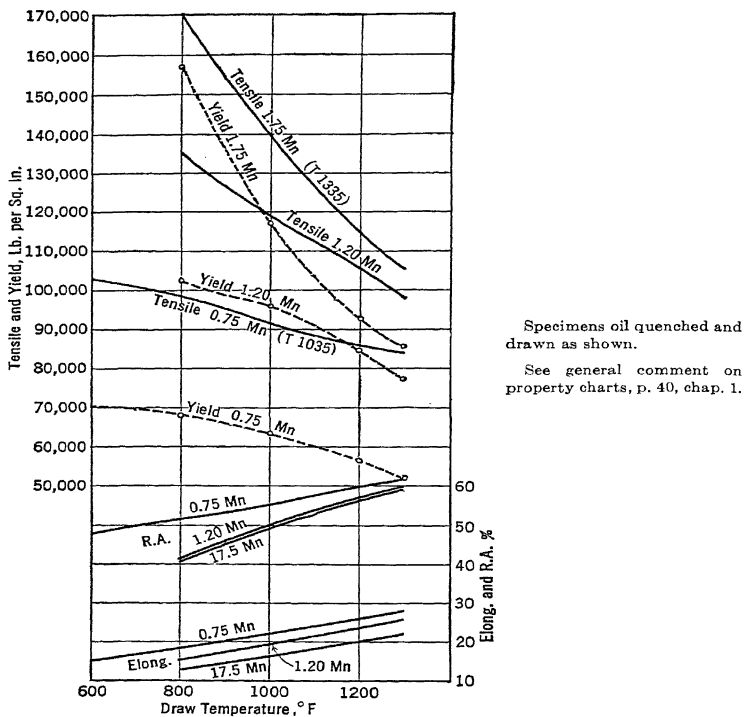


Fig. 4.—Effect of increasing Mn in 0.35% C steels.

In 0.50% C steels held, as in “austempering,” at the temperature where the transformation occurs most rapidly (around 1000°, see Fig. 10, Chap. 6, Vol. I), the delay in the starting of the transformation is

0.5% Mn	under 1 sec.
1.0% Mn	1 sec.
2.0% Mn	10 sec.
4.0% Mn	400 sec.

Whereas 4% Ni only retards it for 2 sec.

Heat-treatment temperatures are lowered by Mn as follows, according to Strauss<sup>3</sup>: 1% Mn lowers  $A_{c3}$   $115^{\circ}$  with very little further

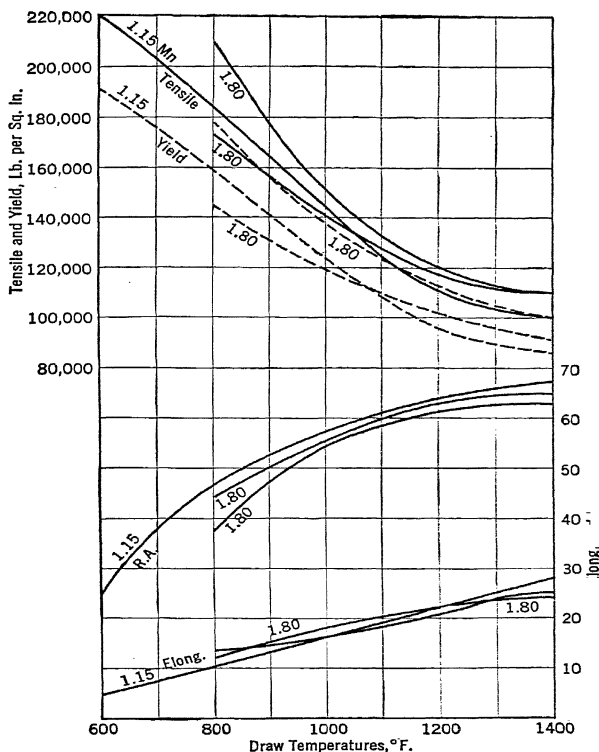


FIG. 5.—Properties of wrought, medium Mn steels, with (a) 0.40 C, 0.22 Si, 1.15% Mn, and (b) 0.37 C, 0.24 Si, 1.80% Mn. Tests on specimens heat treated in 1-in. diam. (a) was first normalized from  $1650^{\circ}$ , then water quenched from  $1500^{\circ}$ . (b) was tested in two lots, one water quenched from  $1450^{\circ}$ , one from  $1550^{\circ}$ . After quenching the steels were drawn as shown. (Bullens.)

See general comment on property charts, p. 40, chap. 1.

decrease with higher Mn up to 3%. The eutectoid C content falls to about 0.78% for 1% Mn and 0.67 for 2%.

**Manganese as Deoxidizer or Alloy.** The dividing line between a plain C steel, with Mn used only for deoxidation and for control of S,

and a Mn steel with Mn present for its alloying effect, is extremely hard to draw. With normal S contents, 0.30% Mn is sufficient to combine with the S and prevent the formation of FeS. Anything above that could be called a Mn alloy steel, for the Mn plays a definite part in conferring physical properties. In cast steels those of less than 1% Mn are considered C steels; those more than 1%, alloy steels; those of 1 to 3% Mn are called "medium Mn." From the pricing point of view, the Mn in wrought steel has to reach 1.50% before the steel takes the "alloy extra." However, the S.A.E. "carbon" steels run up to 0.90% Mn and some of the "X" grades run up to 1.00 and 1.20%. These and the recognized Mn steels (T1300) are listed below:

#### S.A.E. MANGANESE STEELS, X AND T SERIES

##### High-Mn, High-S, Free-Machining Steels

	%C	%Mn	%P	%S	
X1314*	0.10-0.20	1.00-1.30	0.045 max.	0.075-0.15	0.15-0.30
X1315	0.10-0.20	1.30-1.60	0.045 max.	0.075-0.15	(for all
X1330	0.25-0.35	1.35-1.65	0.045 max.	0.075-0.15	here
X1335	0.30-0.40	1.35-1.65	0.045 max.	0.075-0.15	listed)
X1340	0.35-0.45	1.35-1.65	0.045 max.	0.075-0.15	

##### Low-S, Mn Steels, Ordinarily classed as Plain C

X1015	0.10-0.20	0.70-1.00	0.045 max.	0.055 max.
X1020	0.15-0.25	0.70-1.00	0.045 max.	0.055 max.
X1025	0.20-0.30	0.70-1.00	0.045 max.	0.055 max.
X1055	0.50-0.60	0.90-1.20	0.040	0.055 max.
X1065	0.60-0.75	0.90-1.20	0.040	0.055 max.

##### Low-S, Recognized Mn Steels

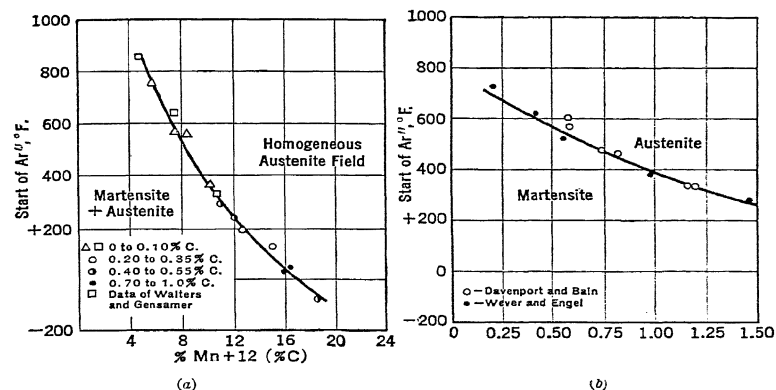
T1330	0.25-0.35	1.60-1.90	0.040 max.	0.050 max.
T1335	0.30-0.40	1.60-1.90	0.040 max.	0.050 max.
T1340	0.35-0.45	1.60-1.90	0.040 max.	0.050 max.
T1345	0.40-0.50	1.60-1.90	0.040 max.	0.050 max.
T1350	0.45-0.55	1.60-1.90	0.040 max.	0.050 max.

\* Cold drawn X1314 in 1-in. rounds, gives 75,000-85,000 tensile, 55,000-65,000 yield, 15-25% elong., 45-55% R.A.

**Balance between Manganese and Carbon.** That so many gradations as to C and Mn content are used and commercially available is eloquent of the value of Mn. Their existence also shows the necessity of careful matching of the C and Mn contents for the particular service in hand. Note that the highest C steel listed, X1065, carries only

slightly over 1% Mn and that the 1.75% Mn series (T1300) stops at 0.45–0.55% C.

In the carburizing grades, raising the Mn content tends to increase the rate and depth of carburization. (See Chap. 14.)



(a) Start of hardening transformation,  $Ar''$ , plotted against equivalent Mn content.

(b) Start of hardening transformation plotted against equivalent C content for C steels with less than 1% Mn.

Fig. 6.—Lowering of  $Ar''$  by combined effect of C and Mn. (Scott and Hoop.<sup>2</sup>)

**Effect of Manganese Content.** Sisco<sup>4</sup> has collected data showing that in quenched and tempered 0.20% C steels the difference between 0.50 and 0.75% Mn is evidenced by distinctly higher hardness, tensile and yield strengths at any given tempering temperature, with some diminution in ductility at the low draws. With the specimens used, however, the Izod impact of the higher Mn steel was equal at the low draws and distinctly superior at the high draws. For comparison, data from the "A.S.M. Handbook"<sup>5</sup> for a  $\frac{3}{4}$ -in. bar of approximately 0.30 C, 1.50 Mn, 0.10% S, oil quenched, are included. At a 930° draw, water- and oil-quenched specimens gave, respectively:

Composition, %C..	0.20	0.20	0.20	0.20	0.30
%Mn.	0.50	0.75	0.50	0.75	1.50
Tensile...	94,000	107,500	Oil {	89,500	116,000
Yield....	71,500	80,500		65,000	92,000
Elong. %.	27	23		29	20
R. A. %..	60	58		61	55
Impact...	19	56		24	65
				60	
				65	

Data from Strauss<sup>3</sup> for specimens from surface and center of water-quenched (from 1550°) 3 $\frac{7}{8}$ -in. rounds of a steel with 0.34 C, 1.88% Mn, drawn at 900°, follow:

	Tensile	Yield	Elong.	Red. Area	Izod F.
Surface.....	141,500	102,000	17.5	55	59
Center.....	115,500	71,500	19.5	56	50

According to Hodge and Sadler,<sup>6</sup> quenched and tempered steels of 0.35–0.48 C, 1.10–1.70% Mn are used in oil well casing for great depths.

**Welding.** We shall comment in Chap. 11 on the air-hardening effect in welding the low-alloy, high yield strength, steels used without heat treatment, pointing out that when such steels are strengthened with Mn it is particularly essential that C be kept very low, or else the steel be joined by riveting, not by welding. When hardening is desired, and is to be followed by tempering, the propensity toward hardening shown by Mn steels is helpful. Böhne<sup>7</sup> suggests for use as welded aircraft fittings, and even for aircraft tubing (usually made of CrMo steel) “Izett” steels of 0.11–0.20 C, 0.50 Si, min., and either 1.00 or 2.00% Mn, min. The term “Izett” usually refers to Al-treated steels of controlled grain size and high coarsening temperature. The 1% Mn steel is credited with 71,000 tensile, 43,000 yield, 16% elong. and the 2% grade with 100,000 tensile, 65,000 yield, 12% elong., which properties are said to be retained after welding.

**Rail Steel.** In all cases there is a limit for the useful amount of Mn, and the amount that can be tolerated decreases as the C increases. An outstanding example of this is in “medium Mn” rail steel, used because of its superior wear-resistance.

The ordinary rail steel specification of 0.69–0.82% C with 0.70–1.00% Mn is replaced by 0.55–0.70% C when 1.20–1.70% Mn is used; but care has to be taken to balance the C and Mn. Either a 0.55 C, 1.70 Mn or a 0.70 C, 1.20% Mn steel makes a good rail, but a rail of 0.70 C, 1.70% Mn is too brittle. With the proper balance, as Thum<sup>8</sup> points out, the Mn rails meet the drop test, and, although the Izod impact is only about 4 ft.-lb., this is double that of a regular rail steel. Early failures of medium-Mn rails were definitely traced to running both C and Mn on the high side.

**Manganese in Free-Cutting Steels.** Starting with low-C steels we meet the free-cutting or screw-stock steels which have been discussed in

Chap. 2, in which a certain degree of hardness and brittleness is required in order that they may not tear on machining and that short chips instead of long curly ones may be produced. The Bessemer screw stocks 1112 and X1112, or their analogs, made by suitable additions to open-hearth steel, harden the steel by high P and introduce very high S—up to 0.20–0.30%, together with high Mn, 0.60–0.90%, in order to produce MnS inclusions that will make the chips break short. These 0.08–0.18% C screw-stock steels are not designed either for carburizing or for heat-treatment.

The free-cutting steels, 1115 and 1120 with normal P and 0.075–0.15% S, that are used for carburizing, in the lower C ranges, are higher in Mn than their 1015 and 1020 counterparts and the X1314 and X1315 carburizing grades carry 1.25 to 1.50% Mn in order to accelerate carburizing and to produce better response of the case to quenching after carburizing.

The free-cutting steels designed for oil quenching and tempering, X1330 to X1340, carry about 1.50% Mn. A steel, not included in the S.A.E. numbering system, of 0.45–0.55 C, 1.00–1.30 Mn, and 0.08% max. S is also available. It will be noted that as the C goes up, the Mn comes down, and the S (hence the MnS inclusions) is also held below the maximum for other free-cutting steels.

**Properties of Free-cutting Steels.** The static properties of these high-S, free-cutting steels are practically indistinguishable from those of the normal, T1330–1340 series with 0.05% max. S. The impact resistance is of course reduced by the inclusions and, when heat treated to high hardnesses, such steels might be looked upon with some suspicion under repeated stress. The relation of the direction of the applied stress to the direction of the elongated plates and stringers of MnS would determine the permissibilities of such steels for a given use. A compromise between the disadvantages of reduced impact resistance and the advantage of improved machinability is often struck. Since machinability enters so greatly into production costs, the free-machining heat-treatable steels are used very widely, especially when a deeper hardening steel than plain C is needed.

**Impact Properties.** When machinability is not so vital, or when a diminution in impact cannot wisely be borne, the normal "T" MnC series finds use.

When water-quenched and tempered at 800° or above, as Hruska<sup>9</sup> shows, for a steel of 0.21 C, 1.69 Mn, 0.10% S, impact tests of the free-cutting Mn steels show fair toughness. Strauss<sup>3</sup> compared two steels, one of 0.38 C, 1.75 Mn, 0.10% S and one of 0.34 C, 1.88 Mn, and less than 0.03% S, after a variety of heat treatments. In these tests the

high-S steel consistently showed as good impact as the low-S. The low-S steel showed more tendency toward temper brittleness on slow cooling from the draw, a fault that has been ascribed to the Mn steels as a class along with NiCr steels, but which Strauss ascribes to melting practice rather than to composition. A Mo addition of about 0.20% avoids the tendency toward temper brittleness in medium-Mn steels as well as adds depth-hardening properties. Greaves and Jones<sup>10</sup> and Burns<sup>11</sup> secured very satisfactory properties, with freedom from temper brittleness, on a steel of 0.30 C, 2.30 Mn, 0.50% Mo, oil hardened, even though this amount of Mn is considered high by most authorities. On the whole the impact resistance of the 1.25–1.75% Mn steels, when drawn back only to 200,000 tensile, is somewhat lower than that of some other alloy steels, but when drawn to lower strengths they should show satisfactory toughness.

**Sensitivity.** In the early literature, the medium-Mn steels were appraised as treacherous, brittle, and sensitive, but, as has been stated

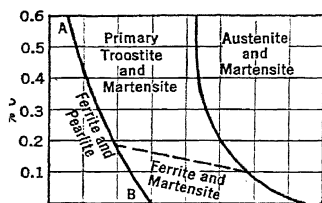


FIG. 7.—Effect of C and Mn content upon structure. Normalized from 1740°. (A.S.M. Handbook.<sup>5</sup>)

in earlier editions of this book and emphasized by Strauss,<sup>3</sup> Hruska,<sup>9</sup> and Thum,<sup>12</sup> this early appraisal is not justified when the proper balance of C and Mn is observed.

**Normalizing.** Strauss points out that with less than 0.30% C useful steels may be obtained with 4% Mn, but, with 0.30–0.50% C, the Mn should be held below 2%. The "A.S.M. Handbook"<sup>5</sup> states that even low-C steels with 1.60% or more Mn tend to show martensitic-appearing structures, and shows Fig. 7. Steels to the right of line AB are air hardening in small sizes. From this it can be seen that the higher C, S.A.E. "T" steels, with the maximum 1.90% Mn allowed, are very close to the border line where erratic results may be obtained because of strong air-hardening propensities.

Austin<sup>7</sup> studied a steel of 0.40 C, 1.40% Mn and one of 0.63 C, 1.80% Mn. The first is in the left-hand zone of Fig. 7, the second is in the next zone.

# NORMALIZED Mn STEELS

Steel		Treatment	Tensile	Yield	Elong. %	R. A. %	Izod
%C	%Mn						
0.40	1.40	As rolled	117,000	68,000	26	36	4.5
0.40	1.40	1470° air	116,000	74,500	28	39	17
0.40	1.40	1470° air	112,000	77,000	34	50	29.5
		1380° air					
0.63	1.80	As rolled	135,000				1.5
0.63	1.80	1470° air	170,000				2
0.63	1.80	1470° air	128,000				8
		drawn at 1150°					

The attempts at normalizing the steel high in C and Mn left it with a pseudo-martensitic structure, and even a high draw after normalizing did not give it good impact resistance.

**Grain Size.** Additions for grain-size control are helpful in improving the normalized structure of the Mn steels. Without such additions the Mn steels tend to have coarse grain size. This, combined with their depth-hardening propensities due to the alloying effect of Mn, gives them an air-hardening tendency, unless the C is kept low, even when the steels are fine grained. In this connection, see Fig. 6, Chap. 10. According to the "Welding Handbook"<sup>13</sup> a steel of 0.17 C, 1.34 Mn, 0.23 Si, 0.15% V hardens, on cooling from a welding heat, to 230–260 Brinell, which is above the 200 Brinell limit set for suitable toughness in welded steel; hence, even such a fine-grained steel is appraised as requiring a stress-relief anneal. Swinden<sup>13a</sup> comments that the addition of 0.37% Ti to a steel of 0.15 C, 0.20 Cu, 1.44% Mn, suppresses weld-hardening.

**Carbon and Manganese Relations.** Thum,<sup>6</sup> Fig. 8, shows the relations between C and Mn in commercial steels.

An exception to the general rule of avoiding the combination of high Mn and high C occurs in steels for shear blades which, as Thum<sup>14</sup> remarks, are sometimes made of steel with 0.80% C and 2% Mn, oil hardened and drawn.

**Manganese in Large Forgings.** The effects of Mn are particularly noticeable in medium-heavy forgings—say, from 4- to 12-in. sections. A 0.40% C steel crankshaft would require a full heat treatment to meet a specification of "80,000–50,000–25–45"; but the same section from a properly forged 0.40% C, 1.10 to 1.30% Mn steel would produce these results by normalizing. Representative results from 0.38 C,



1.08% Mn steel forgings, homogenized, and then normalized from 1475–1500°, are as follows:

Tensile.....	103,500	98,500	92,000
Yield.....	66,000	60,000	51,500
Elong. %.....	24.5	25	27
R. A. %.....	48	47.5	48

The same forgings, quenched and tempered, would, of course, give somewhat higher values, particularly in elongation and reduction of area. With a higher Mn content the C-Mn steels compare favorably with Cr-V steel in static properties, and are not so sensitive as the latter to breakage from cooling stresses directly after forging.

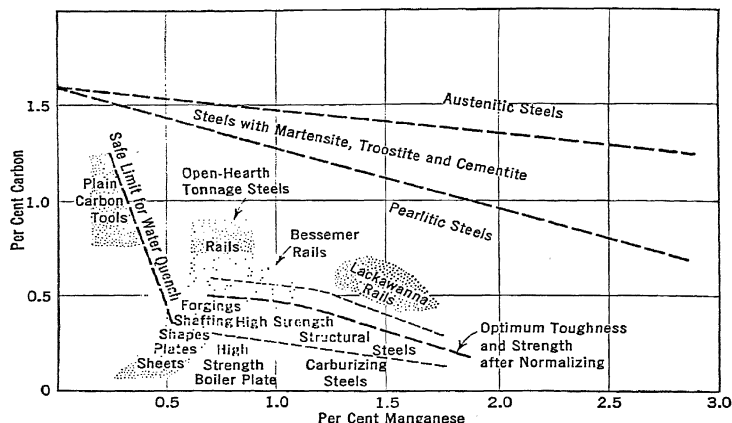


FIG. 8.—Relation between C and Mn in tonnage steels. (Thum.<sup>8</sup>)

**Rifle Barrels.** Rifle barrels present an interesting example of the high combination of strength and toughness obtainable from these steels. They also require an unique machinability. With a steel containing 0.45 to 0.55% C and about 1.25% Mn, the average properties after heat treatment were 115,000 to 125,000 tensile, 80,000 to 90,000 yield, 22 to 24% elong., and about 55 to 60% R. A.

The period of saturation, especially on the draw heat, exerts a tremendous influence upon the results to be obtained from the heat treatment of the pearlitic C-Mn steels.

**Axles.** The group of C-Mn steels analyzing 1.60 to 1.90% Mn with either 0.30 to 0.40 or 0.40 to 0.50% C, according to the recommendation of the individual steel mill, has considerable use for auto-

motive axles. The mills find that they can control very closely, during rolling, the properties of the rolled bar (as contrasted with greater variations in similar straight C steels), making it possible in some cases to omit heat treatment to produce results in 2- to 3-in. diam. bars to meet minimum specifications of 130,000–110,000–15–45 in the 0.30–0.40% C range and some 10,000 lb. higher in the 0.40–0.50% C range.

As was stated in Chap. 13, Vol. I, steel of very high Mn content has been used with the flame-hardening process.

**Non-deforming Steels.** The higher C, medium-Mn, steels retain considerable austenite on oil quenching, hence do not change dimensions on hardening as much as the plain C steels. If the steel is drawn back at 300–400° to about 60 Rockwell C., so that some of the austenite is still retained, the total change in dimensions from the unhardened state is very small. Such “non-deforming” tool and die steels (discussed in more detail in Chap. 22) contain about 0.90% C and 1–1.90% Mn, usually slightly over 1.50% when no other alloys are added, and in that neighborhood when additions are made such as about 0.70% of Cr, or W, 0.30% Mo, 0.25% V, or some combination of these. These carbide formers are added when improved wear resistance is desired. Quenching temperature is about 1450°. Uniformity in quenching temperature is necessary if final dimensions are to be held constant from piece to piece.

That the  $A_{c3}$  temperature is lowered by Mn, and that Mn steels are prone to grain growth, combine to hold the proper quenching temperature down, and to give relatively small leeway in hardening temperature. This fault may be corrected by grain-size control with Al, V, Ti, or other elements that tend to produce high coarsening temperature.

**Vanadium, Aluminum, and Titanium Additions.** The combination of Mn for depth hardening and V for production of fine grain (even on normalizing from fairly high temperature) and, by fineness of grain, for correction of any tendency of the Mn steels towards low impact, utilizes the beneficial effects of both elements. The “C–V” steels are almost invariably on the high side in Mn, and the increased strength shown by such steels is really chiefly ascribable to the Mn; the rôle of the V is to increase toughness.

With better understanding of grain-size control by Al and Ti these elements have been utilized for control, so that the use of the more expensive V has diminished, save in tool steels, which benefit by the hardness of vanadium carbide and which get more generous doses of V than when it is used solely for grain-size effects.

**High-temperature Properties.** Mn, in the "medium Mn" steel range, appears to have a helpful effect on high-temperature properties, though evidence of this is somewhat more striking in short-time, high-temperature tests than in long-continued creep tests. It is not certain whether this effect is actually due to the Mn in itself or to its tendency to produce coarse grain size. Nor is the best range of Mn percentage, and that of other alloying elements, most compatible with it for high-temperature service, clearly worked out. However, for the present at least, a Mn content of say 1.25–1.50% in pearlitic steels for high-temperature service is not avoided and may be helpful.

**Steel Castings—Requirements.** When the minimum properties of grade C cast steel, for railway use, i.e., 90,000 tensile, 60,000 yield, 22% elong., 45% R. A., are sought as well as notch toughness of 20 or 25 ft.-lb. Izod minimum (which is usually not specified but which is nevertheless desired), the choice of alloying elements generally starts from the premise that at least 1 % Mn will be present. The real question usually turns out to be whether the Mn itself shall be raised, to or towards 1.75%, whether other strengthening elements shall be used in addition, and whether V, Ti or Al shall be used for grain-size control.

Owing to the danger of quenching the very large and intricate castings used by the railways, and to the better properties usually obtained by normalizing instead of full annealing, the actual problem in this case, and in many other cases where high-strength steel castings are needed, is the selection of a cheap steel that will respond to normalizing. Single normalizing rather than double and, if possible, the avoidance of necessity for tempering after normalizing are sought in order to reduce production costs.

**Plain Medium Manganese Cast Steel.** A steel of 0.33 C, 0.38 Si, 1.50% Mn (A) was found by Zeuge,<sup>15</sup> one of 0.40 C, 0.40 Si, 1.40% Mn (B) was reported by Lorig and Williams,<sup>16</sup> and one of 0.35 C, 0.40 Si, 1.40% Mn (C) reported by Strauss<sup>3</sup> to show:

Treatment	Tensile	Yield	Elong. %	R. A. %	Charpy Impact
A—3-hr. 1650°, norm. no draw....	104,000	61,000	23.5	49.5	11
A—3 hr. 1650° norm. 3 hr., 1250°.	89,000	51,000	28	58	13
B—1650° norm., no draw.....	95,000	60,000	20	40	13
B—1650° norm., drawn 1250°.....	85,000	52,000	24	45	15
B—1650° water-quenched, drawn 1290°.....	{ 80,000 85,000	{ 50,000 55,000	26–30	50–55	
C—double norm. and drawn.....	108,000	67,500	26	59.5	19.5 Izod

## MANGANESE CAST STEELS

The required properties are approached but the desired combination is not obtained on single normalizing. Even quenching and tempering do not give the desired yield strength when suitable toughness has been obtained. Double normalizing brings everything but the impact up to requirements.

**Effects of Variations in Manganese.** Grotts<sup>17</sup> has reported on a range of Mn contents in double-treated steels. The steels contained about 0.34% C, were first annealed at 1650°, then normalized from 1550°, and finally drawn at 700°. The results were:

%Mn	TENSILE	YIELD	ELONG.	R. A.	Izod
0.80	83,000	39,500	27	40	28
0.92	84,500	41,000	26	39	23
1.12	80,500	41,000	30	47	16
1.34	86,000	45,000	24	50	18
1.53	95,000	54,000	28	59	22
1.68	103,000	55,000	25	51	21
1.76	107,000	55,000	24	54	24
1.85	110,000	63,000	22	44	26

The last one meets the yield requirement but has no margin in ductility. It is evident that some other elements must be called upon to aid the Mn.

**Nickel-Manganese.** Armstrong<sup>18</sup> studied the addition of Ni to medium Mn steels. A representative composition is 0.29 C, 1.25 Mn, 1.25% Ni. This is designated as A in the table below, and was normalized from 1600° and drawn at 1100°. Zima<sup>19</sup> reported on one of 0.33 C, 0.38 Si, 1.20 Mn, 1.06 Ni, normalized from 1600°, drawn at 1200°, designated B. This steel, double normalized from 1750° and 1550°, drawn at 1200°, is designated C. One of 0.33 C, 0.35 Si, 1.10 Mn, 1.25% Ni, normalized from 1650°, drawn at 1250°, from Nickel Alloy Steels,<sup>20</sup> is designated D. Armstrong<sup>21</sup> in a later article gives data for several steels of 0.27–0.31 C, 1.40–1.60 Mn, 1.40–1.60% Ni, normalized 1500–1550°, drawn 950–1200°, whose ranges are entered as E.

STEEL	TENSILE	YIELD	ELONG.	R. A.	Izod
A	98,000	66,500	25	53.5	
B	99,000	63,500	23.5	48	43
C	97,000	65,000	25.5	47.5	36
D	{ 90,000– 95,000	{ 62,500– 65,000	25	40–50	
E	{ 91,000– 102,000	{ 61,000– 70,000	{ 24.5– 28.5	55–64	40–65

These steels are so strongly air hardening that the draw after normalizing cannot be omitted.

**Variations in Composition.** Higher Ni may be used, e.g., 0.30-0.33 C, 1.40-1.60 Mn, and 1.50-1.75% Ni, normalized from 1600° and drawn at 1100° to give about 100,000 tensile, 65,000-70,000 yield, 25-27% elong., 50-60% R.A. so that the margin of safety over the specification requirements is increased as the Ni increases. Decrease of Ni and increase of Mn may, according to Grotts,<sup>17</sup> be carried out according to the following relations for 0.30-0.40% C:

Mn %	Ni %
1.05 — 1.20	1.00
1.35 — 1.55	0.75
1.70 — 2.00	0.60
2.00 — 3.00	0.45

A steel of 0.34 C, 0.51 Si, 1.50 Mn, 0.77% Ni, double normalized 1800°-1550°, drawn at 1280°, gave 98,500 tensile, 65,000 yield, 24% elong., 40% R.A., and Grotts<sup>17</sup> cites one of 0.34 C, 0.44 Si, 1.48 Mn., 0.47% Ni, treatment not stated, with 94,000 tensile, 54,500 yield, 28.5% elong., 58% R.A. The lower Ni steels thus tend to be deficient in either yield or ductility, unless some grain-control addition is made.

The Ni-Mn composition perhaps most used is 1.00-1.25% Mn plus about 1.25% Ni, with or without 0.10-0.15% Mo. The Mo addition enables the use of a high enough draw temperature to get satisfactory ductility without sacrificing yield strength. The impact is also improved.

With suitable heat treatment, the Mn-Ni and, still more, the Mn-Ni-Mo cast steels offer certainty of meeting class C requirements. The cost and the imperative necessity of drawing after normalizing are drawbacks. It will be brought out in Chap. 5 that Cu may serve as a substitute for Ni in a Mn steel.

**Manganese-Molybdenum.** Control of grain size by some addition, in steels for normalizing, is much utilized in wrought steels. It is also effective in cast steels.

Lorig and Williams<sup>18</sup> cite 0.30-0.40 C, 0.30-0.40 Si, 1.60-1.80 Mn, 0.30-0.40% Mo (A), and 0.25-0.35 C, 0.30-0.40 Si, 1.25-1.45 Mn, 0.30-0.40% Mo (B), heat treated as shown below:

	TENSILE	YIELD	ELONG.	R.A.	CHARPY
A 1650° normalized 1250° draw . . .	96,000	68,000	25.5	57.5	23
A 1650° oil-quenched 1200° draw . . .	102,000	76,000	26	58	29
B 1650° normalized 1275° draw . . .	82,000	57,000	27	61	20
B 1650° water-quenched 1200° draw	101,000	81,000	22	59	29

Still higher Mn contents are used in special cases. For wear resistance on sheaves of wire rope hoists, the use of cast Mn-Mo steel of both high C and high Mn content is recorded,<sup>22</sup> for example, 0.45 C, 2.25 Mn, 0.30% Mo. Heavy castings of such compositions are heat-treated to 140,000–165,000 tensile for this use.

Because of the air-hardening tendency conferred by Mo, the Mn-Mo steels, like the Mn-Ni steels, must be drawn after normalizing. Other examples are given below:

MN-MO AND MN-CU-MO

C	Si	Mn	Cu	Mo	Norm.	Draw	Tensile	Yield	Elong. %	R.A. %	Izod
0.25	0.40	1.40	....	0.25	1700°	1200°	93,000	70,000	25	60	25
0.30	0.40	1.40	....	0.40	{ 0710° 1500°	1250°	89,000	70,000	27.5	61	50
0.25	0.40	1.30	....	0.50	1600°	1250°	103,000	83,000	21	51	45
0.30	0.40	1.00	0.90	0.20	{ 1600° 1500°	1150°	95,000	68,000	25	53	

If these Mo-containing steels are not drawn above 1100° the impact tends to be very low. Double normalizing and drawing at 1150–1250° according to composition seem necessary for the best combination of properties, though the tensile tends to drop with this treatment, i.e., the steel shows a high yield ratio.

The Mn-Mo is especially suitable for quenching and tempering.

**Mn-Cr, Mn-Cr-Si, Mn-Cr-Si-V.** Addition of Cr to Mn cast steel has a somewhat similar effect to the addition of Mo, since both elements increase hardenability. Such steels are sensitive to slight changes in C, Mn, and Cr or Mo content so that normal variations in composition may require altering the draw temperature to fit.

C	Si	Mn	Cr	V	Norm.	Draw	Tensile	Yield	Elong. %	R.A. %	Izod
0.37	0.45	1.50	0.50	....	1650°	1250°	117,000	73,000	20	48	
0.25	0.50	1.20	0.50	0.17	1650°	none	107,500	85,000	23	50	
0.32	0.70	1.15	0.50	0.20	1650°	1290°	97,000	71,000	28	60	42
0.30	0.50	1.30	0.65	0.09	1650°	1290°	105,000	80,000	29	60	57

A high draw temperature is required to secure the desired toughness. Presumably Ti could be used for grain refinement instead of V.

**Manganese-Vanadium.** As the preceding has indicated, the addition of about 0.10% V to the medium Mn cast steels (to be treated by single or double normalizing, preferably followed by a draw) is helpful

because of its grain refining effect. Hamilton<sup>28</sup> gives data showing the following ranges:

	No V	V
Tensile.....	91,000 - 112,000	93,000 - 109,000
Yield.....	60,000 - 70,000	69,000 - 86,000
Elong. %.....	25 - 31	26 - 30.5
R.A. %.....	52 - 60	55 - 62
Izod.....	16 - 28	46 - 63

Strauss<sup>3</sup> and Grotts<sup>17</sup> give the following:

C %		Mn %	V %		TENSILE	YIELD	ELONG.	R.A.	Izod
0.35	0.40	1.40	....	(a)	108,500	67,500	26	59.5	19.5
0.35	0.40	1.40	0.10	(a)	103,500	77,500	27.5	57.5	53
0.39	0.31	0.58	0.26		82,500	53,000	25	52	22
0.40	0.32	0.96	0.18		92,000	57,500	26	50.5	32.5
0.32		1.38	0.10	(c)	98,000	70,000	28	57.5	58

(a) Double normalized and tempered (Strauss).

(b) Single normalized and tempered (Strauss).

(c) Annealed 1650°, normalized 1550°, tempered at 700° (Grotts).

Comparison of 3 and 4 shows the benefit of increase in Mn over that of the ordinary C cast steel range, and comparison of 1 and 2 shows the benefit from V. As is indicated above in 4, single normalizing does not give as good impact as double. Steel of 0.30 C, 0.40 Si, 1.55 Mn, 0.10% V single normalized from 1550°, drawn at 1050°, runs about 100,000 tensile, 67,500 yield, 26% elong., 55% R.A., 20 to 45 Izod, average about 30; with double normalizing the Izod would be 50-60. If normalized from 1700° it may give 112,000 tensile, 83,000 yield, 23% elong., 54% R.A., 15 Izod.

**Grain Growth.** Unless a subsequent lower normalizing temperature is used, 1650° seems the upper temperature limit for a single normalize (unless the heating time is much shorter than normal), else the capabilities of the steel for high ductility and impact will not be fully utilized. Although V makes the steel resistant to grain coarsening, long times at 1650° or short times at still more elevated temperature allow grain coarsening with consequent loss of impact. This is consistent with the effect of V in wrought steels, where, though at ordinary normalizing temperatures, the presence of V makes the steel resist coarsening, yet, if it is desired to coarsen the steel to improve machinability, it can be done by raising the temperature high enough. The temperature and time required to homogenize castings of large cross section bring the suitable heating temperature for homogenizing

close to the coarsening temperature of the Mn-V steels. This is not an important drawback on castings of small section.

A good composition for the Mn-V steels is 0.28–0.32 C, 1.55–1.75 Mn, 0.08–0.12% V; double normalizing is advisable for best impact and imperative if grain coarsening has occurred in the first treatment. Tempering after normalizing, at 700°–1000°, is carried out for stress relief without altering the properties materially. If it is desired to soften the steel, tempering is done at 1100–1250°. Tempering appears much less imperative for Mn-V than for Mn-Ni or Mn-Mo.

Since the required amount of V has a cost of the same order as the cost of the 1.25% Ni in the Mn-Ni steels, when it is considered that V in return scrap is lost on remelting while Ni is not, there is not much to choose between the two types of steels.

**Manganese-Titanium.** Cheaper means of grain-size control are being eagerly sought. Much attention is being paid to Ti.

Duma<sup>24</sup> reports on a steel, classed as a C-Ti, but in which the properties are considered to be built up by the Mn and Ni as well, which contained 0.33 C, 0.47 Si, 0.86 Mn, 0.45 Ni, 0.18% Ti. Double normalized from 1650° and 1525°, and drawn at 1100° this gave 91,000 tensile, 62,000 yield, 25% elong., 40% R.A., 21 Izod.

Comstock<sup>25</sup> advocates addition of 5 lb. carbon ferrotitanium per ton to basic open-hearth cast medium-Mn steel. For Ti-treated steels of 0.26–0.31 C, 0.40 Si, 1.39–1.68 Mn, 0.04–0.05% Ti, normalized from 1600° and drawn at 1050°, he secured the following ranges:

Tensile.....	91,000 – 106,000
Yield.....	63,000 – 78,000
Elong. %.....	26 – 31
R. A. %.....	50 – 60
Izod.....	28.5 – 55

The Ti addition prevents the grain coarsening to which the medium-Mn steels are prone and so allows normalizing from 1500 to 1750°. Even 1900° may not coarsen this steel.

The properties of the Mn-Ti steel do not appear to be affected by the 1050° draw, and it will probably require upwards of 1200° to cause appreciable softening. It is possible that this steel may be satisfactory without a draw after normalizing. There seems to be little need for double normalizing as the properties are not markedly different with either single or double normalizing. Any improvement in impact appears to be only of the order of an increase from 35 ft.-lb. on single normalizing to 45 on double normalizing. The maximum impact seems



to be somewhat below the maximum for double normalized Mn-V, but this is compensated for by the marked insensitivity to coarsening, so that the normalizing range is wider.

The grain refinement due to the presence of Ti alters the critical cooling rate and makes the Mn steel less strongly air-hardening. If castings are normalized by pulling a car of them from the furnace and allowing them to cool without separating them, the yield strength may be affected. Where the steel would give 67,000 yield on castings cooled separately, it may fall to 57,000 on full annealing or on too slow cooling. Separating the castings for normalizing, as is required also with Mn-V steel, is good practice for Mn-Ti.

The low cost of the Mn-Ti steel, its properties, and the simplicity of the necessary heat treatment make it attractive. It may not be sufficiently fluid to be as satisfactory from the foundry point of view as one might wish. As will be discussed in Chap. 5, there is some reason to expect that an addition of Cu, which is not expensive enough to sacrifice much of the price advantage of Mn-Ti, may overcome this, and some other advantage may accrue from the presence of Cu.

**Manganese-Aluminum.** Since the Mn-V and Mn-Ti steels are based on a balance of C and Mn, plus the addition of a grain-refining element, it is possible that a good Mn-Al steel might be produced. Little information is available as yet on such steels.

Briggs and Gezelius<sup>23</sup> report the following on a 0.34 C, 0.32 Si, 1.46 Mn, 0.05% Al cast steel for specimens from the center of keel blocks as shown below.

Size	Annealed at	Time	Tensile	Yield	Elong. %	R. A. %	Izod
1-in. square block.....	1650°	1 hr.	99,000	60,750	24	49	17.5
8-in. square block.....	1650°	8 hr.	88,500	52,250	18	31	
8-in. block—8 hr. 1650°,	2050°	5 min.	96,800	60,750	22	41	16.5
then re-annealed.....		1 hr.	99,000	58,500	22	46	14.5

Besides the effect of size, the resistance to grain coarsening due to the Al addition is shown. These specimens were annealed instead of using the more common practice of normalizing, in order to study the size effect without complications due to variations in cooling rate. Better impact would be expected in normalized specimens.

Although it is quite possible that the proper balancing of composition and the proper heat treatment may develop a Mn-Al cast steel

comparable with the other Mn steels discussed above, this has not yet been accomplished.

**Manganese in Low-Alloy Wrought Steels.** In the low-alloy, high-yield-strength steels (Chap. 11) for use without heat treatment, Mn and Si may be used more or less interchangeably; e.g., if we are to have 1.50% of Mn + Si it makes little difference whether we use 0.50 Mn, 1.00% Si or 0.50 Si, 1.00% Mn. The Si-Mn combination is not used quite so interchangeably in heat-treated wrought steels, but the introduction of Si is useful, especially in the higher C spring steels and in steels for punches and shear blades, which go under the name of silico-manganese. Actually, the Mn, while high enough in these commercial steels to exert a slight alloying effect, is not above the range of the so-called plain C steels, to that the properties are due more definitely to the high percentage of Si rather than to the slight increase in Mn. Burns<sup>11</sup> has discussed steels of the usual range of Mn in the "medium Mn" category with large amounts of Si, but these, too, will be dealt with under Si steels.

**Summary.** Manganese is both a ferrite-strengthening and a carbide-forming element, and it divides between the two constituents. It raises the yield ratio. It lowers  $Ac_3$  and lowers the eutectoid C content. It does not form a gamma loop. It markedly decreases the critical cooling rate and makes for depth hardening, i.e., it makes austenite sluggish. It does not materially affect the rate of tempering of martensite and troostite, but probably tends to increase high-temperature strength slightly. It is useful in carburizing steels, and in cast steels. Its drawbacks are brittleness, when too high C and too high Mn accompany each other, a tendency toward temper brittleness in steels slow cooled from the draw, and a tendency toward low grain-coarsening temperature. The first fault can be avoided by proper selection from among the many commercial grades obtainable in which C and Mn are properly balanced; the latter two faults can be avoided by suitable additions of other elements, especially Mo for temper brittleness and V or Ti for grain control.

Manganese can be substituted for part of the alloy content in many steels that would otherwise require a larger amount of a more expensive ferrite strengthener or carbide former. It is the mainstay in the production of alloy cast steels. Its cheapness is not the least of its virtues.

#### BIBLIOGRAPHY

1. EPSTEIN, S., J. H. NEAD, and J. W. HALLEY: Choosing a Composition for Low-Alloy High-Strength Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 120, 1936, pp. 309-345. Also *Metals Technology*, v. 3, 1936, Tech. Publ. 697, 31 pages.

2. SCOTT, H., and J. G. HOOP: The Hardening Transformation in Manganese Steels, *Trans. Am. Soc. Metals*, v. 21, 1933, pp. 233-259.
3. STRAUSS, J.: Medium Carbon Pearlitic Manganese Steels, *Trans. Am. Soc. Steel Treat.*, v. 14, 1928, pp. 1-26.
4. SISCO, F. T.: "The Alloys of Iron and Carbon." Vol. II—Properties. New York, McGraw-Hill Book Co., Inc., 1937, 777 pp.
5. American Society for Metals: "National Metals Handbook," Cleveland, The Society, 1936 ed.
6. HODGE, J. C., and C. R. SADLER: Paper before Western Metal Congress, March, 1938, abstracted in *Steel*, v. 102, April 4, 1938, p. 58.
7. AUSTIN, C. R.: Observations on the Effect of Normalizing Medium Manganese Steels on the Microstructure and Physical Properties, *Trans. Am. Soc. Metals*, v. 21, 1933, pp. 435-450.
8. THUM, E. E.: Discussion, *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 455-458.
9. HRUSKA, J. H.: Study of the Pearlitic Manganese Steels, *Heat Treat. Forg.*, v. 16, 1930, pp. 455-459.
10. GREAVES, R. H., and J. A. JONES: Temper-Brittleness of Steel; Susceptibility to Temper-Brittleness in Relation to Chemical Composition, *J. Iron Steel Inst.*, v. 111, 1925, pp. 231-264.
11. BURNS, G.: The Properties of Some Silico-Manganese Steels, *Ibid.*, v. 125, 1932, p. 363-391.
12. THUM, E. E.: The New Manganese Alloy Steels, *Proc. Am. Soc. Test. Mat.*, v. 30, part 2, 1930, pp. 215-240.
13. American Welding Society: "Welding Handbook." New York, The Society, 1938, 1210 pages.
- 13a. SWINDEN, T.: Special Steels and Their Application to Engineering and Shipbuilding, *Metallurgia*, v. 17, March, 1938, pp. 181-185.
14. THUM, E. E.: Heat Treated Low-Manganese Steels, *Iron Age*, v. 123, June 20, 1929, pp. 1691-1695.
15. ZEUGE, D.: Alloy Cast Steels, *Trans. Am. Foundrymen's Assoc.*, v. 37, 1929, pp. 361-384.
16. LORIG, C. H., and C. E. WILLIAMS: Physical and Mechanical Properties of Some Well-Known Cast Steels, A.F.A.-A.S.T.M. "Symposium on Cast Steels," 1932, pp. 76-158.
17. GROTTIS, F.: Representative Properties of Cast Medium Pearlitic Steels, *Ibid.*, 1932, pp. 159-174.
18. ARMSTRONG, T. B.: Properties of Some Cast Alloy Steels, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 286-318.
19. ZIMA, A. G.: Properties and Uses of Some Cast Nickel Alloy Steels, *Trans. Am. Foundrymen's Assoc.*, v. 41, 1933, pp. 199-224.
20. FRENCH, H. J., and J. W. SANDS (editors): "Nickel Alloy Steels." New York, International Nickel Co., Inc., 1934, 7 sections.
21. ARMSTRONG, T. N.: Nickel Steel Castings in Railroad Rolling Stock, *Metal Progress*, v. 33, February, 1938, pp. 163-166.
22. Heavier Loads and High Speeds in Oil Production Lead to Use of Moly Steels, *The Moly Matrix*, v. 4, February, 1937, pp. 1-2.
23. HAMILTON, W. C.: Additions of Vanadium Improves Properties of Medium Manganese Steels, *Iron Age*, v. 129, 1932, pp. 546-548.

24. DUMA, J. A.: Effect of Titanium on Some Cast Ferrous and Nonferrous Metals, *Trans. Am. Soc. Metals*, v. 25, 1937, pp. 788-825.
25. COMSTOCK, G. F.: Titanium Improves Pearlitic Manganese Steel, *Metals & Alloys*, v. 8, 1937, p. 148.  
COMSTOCK, G. F.: and C. L. CLARKE: Effect of Titanium on Some Properties of 17.5 Chromium Steel, *Ibid.*, v. 8, 1937, pp. 42-46.
26. BRIGGS, C. W., and GEZELIUS, R. A.: The Effect of Mass upon the Mechanical Properties of Cast Steel, *Trans. Am. Soc. Metals*, v. 26, 1938, pp. 367-407.
27. Bethlehem Steel Company; "Bethlehem Alloy and Special Steels," Bethlehem, Pa., The Author, 1935, 375 pages.

## CHAPTER 4

### SILICON STEELS

Silicon in amounts up to 0.20% is used in practically all commercial killed wrought steels, and up to 0.40% or more in steel castings, on account of its deoxidizing action. Above these amounts it may be considered as an alloy.

The chief application of Si in commercial steels is in the following:

1. Straight Si steels—nearly carbonless steels with up to 4% or more Si, for use in generators, motors, transformers, and other electric apparatus.

2. High-yield-strength, low-alloy steels. (See Chap. 11.)

3. Structural Si steels, often with Mn on the high side, and often termed "silico-manganese" steel, e.g., with medium-C content and about 1.5% Si and 0.75% Mn, for use in heat-treated gears, springs, etc. Other alloying elements may be added to such a base, and this is resorted to when such a steel is used for tough tools. (See Chaps. 16 and 22.)

4. Silico-chromium steels—a division of the "stainless" steels. (See Chap. 20.)

5. Graphitic steels.

**Silicon Steels.** Like Mn, Si appears in commercial steels, so that its presence in scrap does not contaminate a plain C steel melt, or seriously complicate the melting process. It is another of the cheaper alloying elements. Its effect on yield strength in low-C steels, normalized, is shown in Fig. 1.<sup>1</sup>

**Effect on Critical Temperatures.** Whereas Mn lowers the critical temperature on heating, Si raises it. For example, in a 0.10 C steel, 1.25% Si raises  $A_{c3}$  to 1770°, and 4% to 2000°. In a 0.40 C steel, 2.35% Si raises it to 1650°, 3.65% to 1740°. Silicon forms a "gamma loop" (see Fig. 2)<sup>2</sup> analogous to that produced by P, the extent of which varies with the C content. With 0.10 C and over 3% Si, the steel remains ferritic, not passing through the gamma stage at all; with 0.30 C, about 7.5% Si can be present before the possibility of forming

gamma iron is lost, but in a carbonless Fe-Si alloy it is lost at about 2% Si.

**Carburization and Decarburization.** The increase in necessary heating temperature for normalizing or quenching, as Si is raised, accentuates the chemical activity of the surface; and this plays some part in the fact that the high-Si steels are very prone toward decarburization. Increase in Si content tends to decrease scaling when heating in an oxidizing atmosphere, but decarburization can occur under what little scale is formed. This proneness to decarburize is a distinct drawback of Si steels. It can be avoided by the use of suit-

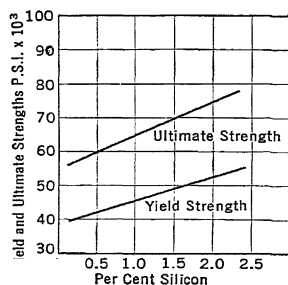


FIG. 1.—Effect of Si on tensile and yield strength of low C steel. (Epstein, Nead and Halley<sup>1</sup> after Paglianti.)

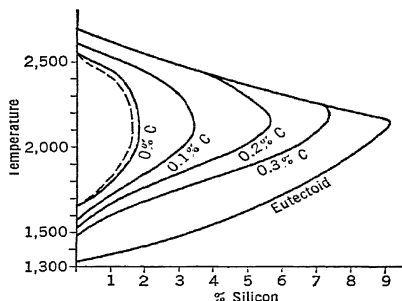


FIG. 2.—Gamma loop in Si steels of varying C content. (After Scheil.<sup>2</sup>) Gamma Fe within the loop, alpha to right of loop. Each loop will have a small area within it, in which gamma and alpha Fe both appear. In this figure this area is indicated only in the case of 0% C.

able “controlled” atmospheres. (See Chap. 12, Vol. I.) In contradistinction to easy decarburization, steels with high Si are carburized only with the greatest difficulty. (See Chap. 14.) This is in line with its graphitizing effect in cast iron.

**Mill Behavior.** The high Si steels are hard to handle in the mill so as to secure a good surface. They pipe heavily so that a large crop discard has to be made. For these reasons they are not used unless their special properties, or their cheapness, give them the call over competing steels.

**General Effects of Silicon.** Silicon does not form carbides in steel; it enters into the ferrite only. It is a recognized ferrite strengthener, but has no hardening effect due to carbides.

Silicon confers some degree of depth hardening, but is not so effective in this respect as is Mn. There is some danger of quench cracking in water hardening of high-C, high-Si steels, e.g., 0.50 C, 2% Si, so that such steels are chiefly used in sections that can be fully hardened in oil, unless, as may be the case in large punches, chisels and shear blades, an oil quench will not give sufficient hardness. Silicon steels are not so prone to grain coarsening on heating above  $A_{c_3}$  as are steels of equivalent Mn content. Silicon is a moderate grain growth restrainer, in compositions within the gamma loop. The very low-C, high-Si electric steels, however, have large grain size, as usually prepared.

Silicon does not confer noteworthy high-temperature strength. The special rôle of Si in quenched and tempered steels is to secure cheaply a good degree of toughness in very hard steels, hence such steels are generally used at a hardness more or less approximating "spring temper."

**Straight Silicon Steels, Low Carbon.** The most important use for straight Si steels is for transformer manufacture and for other electric purposes demanding a high magnetic permeability and low hysteresis loss. Hadfield's Si steel, containing approximately 2.75% Si, with C, Mn, and other impurities as low as possible, is representative of this class. His treatment of this steel consists of first heating to about 1950° and cooling quickly, and then heating to 1380° and cooling very slowly, this treatment being followed, in some cases, by reheating to 1475° and cooling very slowly.

Another Si steel, used in place of dynamo sheet iron, specifies similar C, Mn, etc., but with a Si content of about 3.25%. The thermal treatment recommended for this steel is a thorough heating at about 1430°–1475°, followed by very slow cooling.

There are a number of other grades for various electric uses; transformer steel may contain 4% or more Si and many fine gradations in annealing practice can be employed to control grain size, upon which the magnetic properties are in part dependent. An irritating fact is that with high-Si content and the desired large grain size the sheets (these steels are usually employed in sheet form) are brittle and tend to show ragged edges on shearing, unless sheared at elevated temperature.

These grades are not used for structural purposes so their mechanical properties are of little interest save as they affect fabrication difficulties. The detailed heat treatment of the various grades of electric sheet is too specific a matter for discussion here. The interested reader is referred to the summary in "The Alloys of Iron and Silicon."<sup>3</sup>

**Quenched and Tempered Silicon Steels.** Kinzel <sup>4</sup> studied the suitability for water quenching of steels of 0.25–0.40% C, with Si varying up to 1.6% and Mn to 1.6%, and came to the conclusion that 0.30–0.35% C with 1% each of Si and Mn was the most promising. Such a steel, water quenched from 1650°, drawn at a range of temperatures, gave:

TENSILE	YIELD	ELONG. %	R. A. %	Izod
210,000	190,000	10	41	23
169,000	153,000	13	48	42
161,000	148,000	15	50	44
135,000	122,000	20	58	72

On account of the limitations of a water quench, Kinzel added Cr to give better depth hardening and thus make the steel more suitable for oil quenching. A steel of 0.34 C, 0.91 Si, 1.04 Mn, 0.95% Cr, oil quenched from 1650°, gave:

	TENSILE	YIELD	ELONG. %	R. A. %	Izod
800° draw. . . .	201,000	178,000	10	38	22
930° draw. . . .	167,000	151,000	15	48	24

Burns <sup>5</sup> explored the steels of 0.40–0.50% C, over a wide range of Mn and Si. He concluded that 0.45% C was as high as it was safe to go with water quenching, that, with Mn above 1%, addition of more than 1% Si injures ductility and impact, and that depth hardening was improved much more by increasing Mn than by increasing Si.

**Spring Steel.** British Si-Mn spring steel specifications call for 0.50–0.60 C, 1.20–2.00 Si, 0.60–1.00% Mn. Burns points out that with Mn and Si both on the low side the steel has inferior hardenability. In deciding which it is better to raise, Si or Mn, he found, from fatigue tests, that mechanical defects on the surface were more harmful with higher Mn, leading to the choice of higher Si, but that trouble from decarburization is accentuated as Si rises, leading to the opposite conclusion.

Actual experience has led to the use of the higher Si type, shown by S.A.E. 9255 and 9260, in the range 0.50–0.65 C, with 0.60–0.90 Mn and 1.50–2.20% Si, together with the use of precautions to avoid decarburization. On oil quenching from 1575 to 1650° in small sizes and drawing at 900° a 0.55 C, 0.70 Mn, 2% Si steel will have a tensile strength of about 205,000, 185,000 yield, with 12% elong. and 27% R.A., according to "Bethlehem Alloy Steels," <sup>6</sup> whereas a plain 0.95% C spring steel, water quenched and drawn at 800°, giving about the



same tensile, will show only about 145,000 yield, 12% elong., 32% R.A. The high yield strength of the Si steel fits it for spring service, while the toughness is not much lower than that of the C steel. At high draws the advantage decreases, but is still in evidence, as Fig. 3<sup>6</sup> shows. According to Colbeck and Hanson<sup>7</sup> and to Mathews,<sup>8</sup> somewhat lower C can be used if the hardening temperature is increased. Mathews studied a 0.47 C, 0.70 Mn, 1.86% Si steel, forged at temperatures varying from 1700 to 2400°, then annealed at 1500° for 2 hr., water quenched from 1750°, drawn ½ hr. at 800°. It gave similar

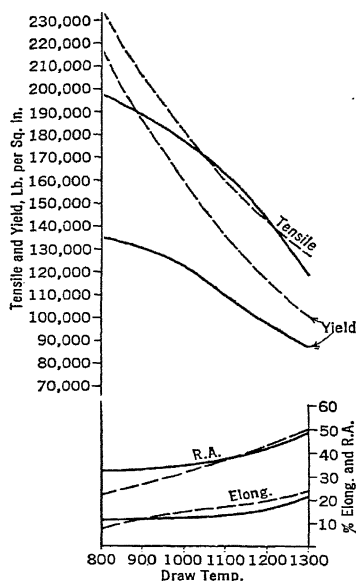


FIG. 3.—Properties of spring steels. Solid lines, C steel, 0.95 C, 0.25 Si, 0.35% Mn (No. 1095), 1475° oil. Dashed lines Si, Mn steel, 0.55 C, 2.00 Si, 0.70% Mn (No. 9255), 1650° oil. Both were first normalized from 1650°. Bars 1-in. round, drawn as shown. (Bethlehem Alloy Steels.<sup>6</sup>)

See general comment on property charts, p. 40, chap. 1.

results regardless of forging temperature, i.e., 248,000 tensile, 227,000 yield, 11% elong., 50% R.A., 425 Brinell. Ability to resist overheating was clearly shown. The composition is at the upper end of those suitable for water quenching. Although the Si-Mn spring steels are often termed brittle, Mathews points out that with proper heat treatment they have much higher impact resistance than high-C spring steel. Quenching from a sufficiently high temperature is necessary to develop best properties. (Compare Chap. 16.)

**Tools.** Under proper control of heat treatment the cheapness, high yield strength and toughness of the Si steels recommend them for some types of gears and in the 400-550 Brinell range especially for punches and chisels. Gill<sup>9</sup> says that such steel will outwear 1% C tool steels three times in punches and shears. When used for tools the steel is usually water quenched if the piece will allow. (Compare Chap. 21.)

Small additions of 0.20-0.35% Cr; of 0.15-0.30% V; or of about 0.50% Mo are made to the Si tool steel base, the C and Si then usually being somewhat reduced from the usual 0.55% C and 2% Si. The addition of one or more of these carbide-formers will confer added hardness, and either V or Mo helps in preventing grain growth. For extreme toughness the Si is dropped to about 1%, the Mn held to 0.35-0.60%, the C held not far above 0.50% and 0.50% Mo added, sometimes 0.30% Cr also. Palmer<sup>10</sup> rates such a steel, water quenched and drawn back to 59 Rockwell "C," as "the strongest and toughest tool steel known." He comments that its propensity toward decarburization is "the only bad habit this steel has." Some makers hold the C a bit below 0.50% but hold the Si about 1.70% in this Si-Mo tool steel.

**Nickel in Silicon Steels.** The addition of another ferrite strengthener, Ni, to the Si steel base increases the toughness at very high strength. A steel of 0.46 C, 1.42 Si, 0.80 Mn, 3.00% Ni, oil quenched from 1575°, drawn at 650°, showed 282,000 tensile, 241,500 yield, 6.5% elong., 24.5% R.A.; and with slightly lower C, Si up to 2.50%, and 0.70% Mo added, the 3% Ni steel also gave good results.<sup>11</sup> Such steels were used for light armor plate during the war. For such special purposes the addition of Ni and Mo would probably be justified, but the increase due to the added alloys is too small to make the steel generally attractive.

Harrison<sup>12</sup> studied the addition of up to 0.90% Si to 0.35-0.45% C steels containing 4% Ni, and found only minor effects. The possibility of replacing Ni by Si, i.e., reducing Ni when Si is added instead of adding it to a steel already high in Ni, seems not to have been fully explored. Since both elements are ferrite strengtheners, such a study might have interesting results. The Ni-Si steels may hold possibilities for resisting certain types of corrosion.

**Silicon Steel Castings.** Since Si in amount of 0.40-0.50% is a normal constituent of cast steel, increasing the amount of Si would be a simple way of making an alloy-cast steel. Schulz and Bonsmann<sup>13</sup> and Körber and Pomp<sup>14</sup> studied such steels in the lower ranges of C content and found that tensile and yield strengths equivalent to those of 0.40% C cast steel could be obtained, with better ductility. Rys<sup>15</sup>

studied higher C contents. The steels of course require a high annealing or normalizing temperature. Rys reported the following:

% C	% Si	% Mn		TENSILE	YIELD	ELONG. %	R. A %
0.20	0.84	0.90	Annealed 1705°	75,000	40,000	25.5	40
0.17	1.11	1.11	Norm. 1830°	75,000	51,000	25	52
0.18	1.22	1.30	Norm. 1830°	84,000	57,000	26	38
0.40	1.50	0.50	X	110,000	68,000	18	30
...	1.50	0.50	X	79,000	47,000	27	45

(X) Heat treatment not stated, probably annealed.

Rys favors "lower C," probably about 0.20%, in the 1.50 Si-0.50% Mn steel, which he says gives the properties in the last line of the table. Just as in wrought steels, Si and Mn may replace each other in the "silico-manganese" type, and as the sum of Si + Mn rises, the C needs to be decreased. Finlayson's <sup>16</sup> low-C, high-Si and Mn cast steels, plus Cu, will be discussed in the next chapter.

**Graphitic Steels.** The presence of Si tends to make cementite unstable, i.e., to form graphite. With the combination of high C and high Si in cast iron, flaky graphite appears in the metal as cast. By reduction in the amount of C and Si, say to 2.20% C, 1% Si, the separation of graphite is avoided, and "white iron" produced. If white Fe, consisting of austenite and free cementite, is annealed in the austenite range, say at 1600°, for an extended period, the excess cementite beyond that soluble in austenite at the temperature used decomposes into Fe and very finely divided (not flaky) graphite or "temper C," peppered through the austenite.

This is the first stage in the production of malleable Fe from white Fe. By very slow cooling through and below the critical range the remainder of the cementite dissolved in austenite, as well as that in the resultant pearlite, can be in turn decomposed to Fe and C so that completely annealed, "long cycle" malleable may consist solely of ferrite, carrying the Si in solid solution, plus separated temper C. The C separating in the second stage tends to form nodules by growth upon nuclei formed in the first stage.

More rapid cooling through the critical produces a pearlitic matrix so that "short cycle" or pearlitic malleable has a high-C, Si steel matrix interspersed with particles of temper C in less nodular form. This high-C matrix is heat treatable.

Alloys with fine temper C particles have extremely good machinability. The presence of these graphitic particles makes the alloys somewhat similar to cast iron in freedom from galling and for service under rubbing conditions with scanty lubrication, possibly owing to

## GRAPHITIC STEELS

91

TABLE I

SLOW-COOLED AT 100° PER HOUR FROM 1500°

OIL QUENCHED FROM 1500°. DRAWN 1100°.

C	Si	Mn	Ni	Cu	Mo	Tensile	Yield	Elong.	R. A.	Charpy	Tensile	Yield	Elong.	R. A.	Charpy
1.50	1.0	0.40	.....	.....	.....	94,000	47,500	19.5	30	.....	154,000	116,000	14	30	(Water quench)
1.20	1.1	0.40	.....	.....	.....	106,500	70,000	27	48	10.5	186,000	112,500	13	31	7
1.40	1.1	0.40	.....	.....	.....	78,250	44,000	24	37	11	175,000	107,500	11	23	4.5
1.75	1.1	0.40	.....	.....	.....	75,000	38,000	25	41	12.5	163,500	105,000	12.5	29	12
1.15	1.05	0.40	0.40	0.40	.....	112,000	77,500	23.5	40.5	10	185,500	121,000	11.5	22.5	5.5
1.15	1.1	0.40	0.75	.....	.....	114,500	80,000	25.5	46	10.5	198,000	125,000	12	29	6
1.15	1.1	0.40	1.50	.....	.....	119,000	86,500	25	46.5	8.5	180,000	142,500	10.5	16	6
1.20	1.10	0.40	.....	.....	0.27	111,500	70,000	25.5	48	10	195,000	146,000	9	20	6.5

Data for the first steel are from ~~our~~ <sup>our</sup> investigations at Battelle Memorial Institute, for the others, from ~~our~~ <sup>our</sup> investigations at Battelle Memorial Institute.

the film of graphite supplied by the particles, possibly to the tiny pockets acting as oil reservoirs when their graphite has been partly or wholly removed.

The phenomenon of graphitization is not unknown in high-C, low-Si steels. When an attempt is made to run the C in tool steels up over 1.20% with even as low as 0.10–0.30% Si, graphitization may occur on annealing, even below  $A_1$ . Kinzel and Moore<sup>17</sup> even report graphitization in a 0.15 C, 0.20% Si steel held just below  $A_1$  for 3

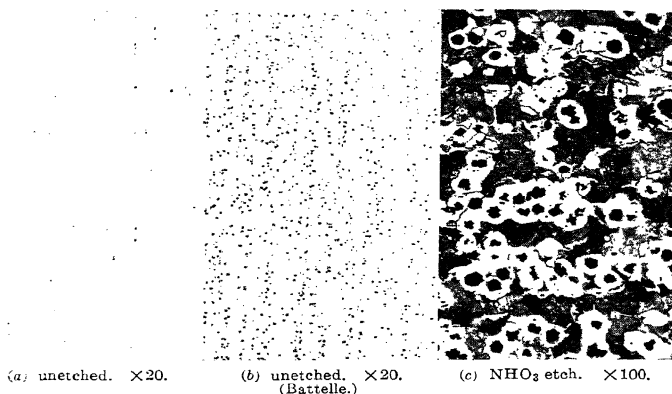


FIG. 4.—Graphitic steel. C, 1.24; Mn, 0.42; Si, 1.39%. (a) is oil quenched from 1800°. (b) and (c) are slow cooled from 1800°. Very little graphitization has occurred in the quenched specimen. In the slow cooled steel much of the carbide in the pearlite has graphitized, producing uniformly distributed temper carbon, but graphitization is not complete.

years. As the Si is increased the tendency toward graphitization increases. Even very small amounts of Al exert a similar tendency.

In the past such attention as has been given to graphitization in true steels has been in the attempt to avoid it. Recently it was appreciated that utilization of the graphitization phenomenon might allow the production of a strong easily machinable steel that could be forged and rolled instead of having to be used in cast forms like cast and malleable irons, especially for wear-resistant uses. Taking a leaf from their experience with malleable, Schwartz<sup>18</sup> and Edmunds<sup>19</sup> suggested such steels, and further development has been carried on by Bonté<sup>20</sup> and others. The steels, according to Schwartz and Edmunds, preferably contain such combinations as 0.35 C, 1.60 Si, 0.30% Mn, or 1.15 C, 1.25 Si, 0.85% Mn. In order to prevent separa-

tion of graphite in freezing and cooling on casting the ingot, 2% Si with 1.50% C or 4% Si with 1.30% C should not be exceeded, and useful properties are obtained with 1% each of C and Si.

The resistance to graphitization of a properly balanced steel, until graphitization is intentionally produced, is shown in Fig. 4a for a 1.24 C, 1.39 Si, 0.42% Mn steel, with only residual traces of other alloying elements, quenched in oil from 1800°. Slow cooling (Fig. 4b) produces uniform temper C distribution. These figures are at X20 to

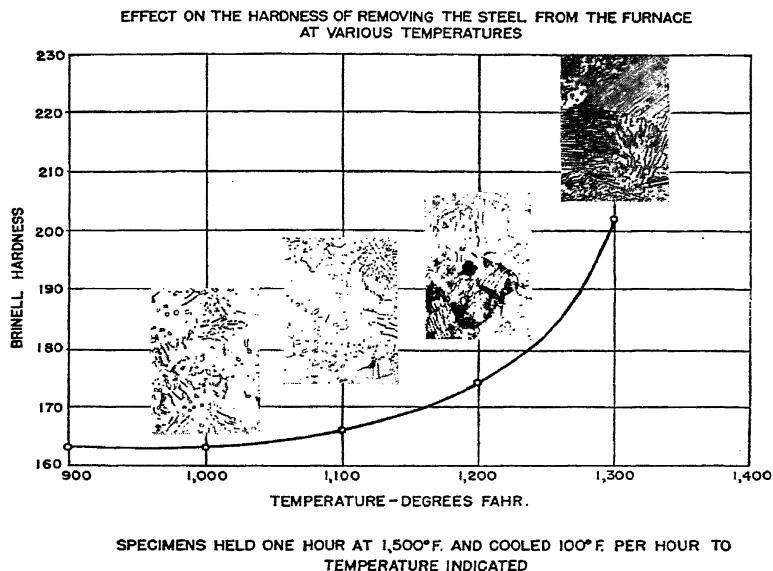


Fig. 5.—Progressive breakup of pearlite areas as slow cooling progresses. This is the same steel as in Fig. 4. Micrographs at  $\times 1000$  reduced to  $\times 500$  in reproduction. (Battelle.)

show distribution. The etched structure at X100 of the slow-cooled specimen reveals the temper C (Fig. 4c). The effect of air cooling from different temperatures after slow cooling, i.e., interrupting the slow-cooling process, is shown in Fig. 5, which follows the break-up of the cementite in the pearlite. The amount of combined C and hence the hardness may be thus regulated. Figure 6 shows both the retained pearlite, and the nodules with their surrounding ferrite. This is a larger view of the micrograph at the right in Fig. 5, for material

whose slow cooling is interrupted at  $1300^{\circ}$ . Structure on slow cooling of a steel of 1.15 C, 1.70% Si is shown in Fig. 7 at X100. The uniform size and distribution of the temper C is notable.

Upon reheating the graphitized steel for quenching, some of the precipitated C can be taken into solution. By adjusting temperatures and times of heating, the matrix can be given different degrees of effective C content.



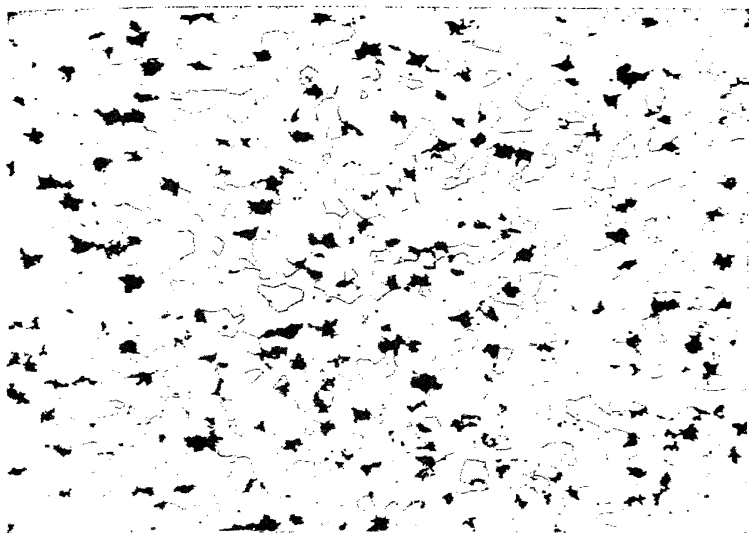
Nitric acid etch  $\times 1000$ . (Battelle.)

Fig. 6.—The same graphitic steel as in Figs. 4 and 5, slow cooled from  $1500^{\circ}$  to  $1300^{\circ}$ , air cooled from  $1300^{\circ}$ . Undecomposed pearlite areas and nodules of graphite with surrounding ferrite are shown.

Depth-hardening properties were studied by oil quenching a disc 10 in. in diam. and  $1\frac{1}{4}$  in. thick from  $1500^{\circ}$ . By splitting the disc flatwise and taking hardness readings along a radius on the cut face, the data of Fig. 8 were obtained.

Bonté reports interesting results from such steels used as drawing dies, the hole being internally quenched from  $1500^{\circ}$  by water spray and drawn at  $300^{\circ}$  to Rockwell C 62–63, the body being 20–22. Remarkable freedom from scuffing of the die was reported; in one case

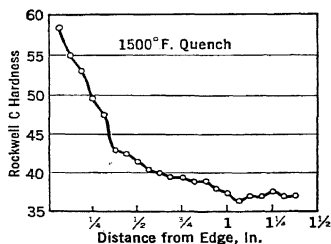
the service given was 31 times that of the die steel previously used. The addition of 0.25% Mo and the reduction of Si to 0.75% are advo-



Nitric acid etch  $\times 100$ ; (Battelle.)

Fig. 7.—Complete graphitization on slow cooling from  $1700^{\circ}$  at  $100^{\circ}$  per hr. of steel of C, 1.15; Mn, 0.37; Si, 1.70%. This completely graphitized steel has a Brinell hardness of 131.

cated by Bonté. He reports that a steel of 1.45 C, 0.35 Mn, 0.75 Si, 0.25% Mo, oil quenched from  $1450^{\circ}$  and tempered at  $300^{\circ}$  to Rockwell C 61–62, used as punches, out-performed an oil-hardening tool steel



C.....	1.17%
Si.....	1.18%
Mn.....	0.39%

Oil-quenched, not drawn. (Battelle.)

Fig. 8.—Depth hardening in graphitic steel.



previously used, by 11 times. The analogy of the matrix of such a steel to the tough oil-hardening tool steel of 0.50 C, 0.40 Mn, 1.00 Si, 0.50% Mo previously mentioned is evident. The temper C particles may have a cushioning effect and thus add to the effective toughness in service.

Bonté<sup>21</sup> finds 0.20–0.50% Mo to act as a graphitizer, rather than as a carbide stabilizer. He gives the following for steels to which has been added Mo, or Mo and Ni.

STEEL	C	Si	Mn	Mo	Ni%		
C	1.45	0.90	0.40	0.25	....	graphitized at 1550°, then quenched, and drawn as show	
D	1.50	0.90	0.40	0.20	1.25		
Steel	Draw		Tensile		Yield	Elong.	Brinell
C	900°		218,000		177,000	8.5	444
C	1100°		164,000		136,000	13	321
D	900°		214,000		176,000	16	415
D	1100°		163,000		134,000	27	321

As Table I shows, use of other alloying elements that favor graphitization, like Ni and Cu, allows a wide range of properties. The strong carbide-former Cr is avoided because of too great stabilization of cementite and prevention of graphitization, but Mo does not interfere, and adds hardenability. The wear-resistant properties, the wide range of hardness obtainable between the quenched and the fully annealed condition, and the ready machinability in the graphitized state make this new family of steels highly interesting. Their applications have not been fully explored, but it seems evident that much more will be heard of the graphitic Si steels.

**Silico-Manganese Steels.** The manufacture of Si-Mn steels in the open hearth must be carefully watched because of their great tendency to piping and segregation, and a large discard must be made in the cropping of the ingots, to ensure good steel. Silico-Mn steels are widely used for automobile springs, both leaf and coil (for "knee action"). Although the lower cost favors their use, their sensitiveness to heat treatment and rather low resistance to shock somewhat limit their field of usefulness. When handled with care, the Si-Mn (and also the Si-Cr) steels give good results in works well equipped for obtaining accurate results in their heat-treatment operations. The temperature limits for quenching are narrower than for most alloy steels, and the steel responds altogether too quickly to variations in heating and cooling. Although these steels will give high static test results upon suitable treatment, the sensitiveness which is inherent in this type of steel when coarse grained may prove a governing

factor. They are no more sensitive, however, probably less so, than plain high-C spring steels.

An analysis for Si-Mn steel for gears is as follows:

C, %	0.43 to 0.53
Mn, %	0.50 to 0.70
Si, %	1.25 to 1.50

Upon suitable treatment, usually quenching in oil from about 1550° to 1650°, followed by tempering to suit the requirements, this gives the following representative results:

Tensile	195,000 to 230,000
Yield	175,000 to 220,000
Elong. %	12 to 8

Silico-Mn spring steels will be discussed in Chap. 16.

**Summary.** Silicon is wholly a ferrite strengthener; it does not enter the carbide, but it tends to promote graphitization of carbide. It interferes with carburization. As a ferrite strengthener it may be used more or less interchangeably with Mn, and thus appears in some of the low-alloy high-yield-strength steels.

It raises  $Ac_3$ , and forms a gamma loop. It is a mild depth-hardening element. It tends to raise the coarsening temperature, thus allowing leeway in heating for hardening; but, since it materially raises the necessary heating temperature, the higher temperature leads to severe decarburization, one of the worst drawbacks of Si in spring steel, unless prevented by the use of a suitable controlled atmosphere.

In rather large amounts it minimizes scaling at high temperatures and for this reason is added to some special, complex, steels; but it does not impart high-temperature strength.

A specific effect is the magnetic properties conferred on very low-C steels. Another is the ability for graphitization it imparts to very high-C steels. More extended treatment of Si steels will be found in "The Alloys of Iron and Silicon."<sup>2</sup>

## BIBLIOGRAPHY

1. EPSTEIN, S., J. H. NEAD, and J. W. HALLEY: Choosing a Composition for Low-Alloy High-Strength Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 120, 1936, pp. 309-345. Also: *Metals Technology*, v. 3, 1936, Tech Publ. 697, 31 pp.
2. SCHEIL, E.: Beitrag zum System Eisen-Kohlenstoff Silizium (Contribution to the System Iron-Carbon-Silicon), *Stahl u. Eisen*, v. 50, 1930, pp. 1725-1727.
3. GREINER, E. S., J. S. MARSH, and B. STOUTON: "The Alloys of Iron and Silicon." New York, McGraw-Hill Book Co., 1933, 457 pages.

4. KINZEL, A. B.: Silicon-Manganese Steels with Chromium Additions for Engineering Applications, *Trans. Am. Soc. Steel Treat.*, v. 14, 1928, pp. 866-876.
5. BURNS, G.: The Properties of Some Silico-Manganese Steels, *J. Iron Steel Inst.*, v. 125, 1932, pp. 363-391.
6. Bethlehem Steel Co.: "Bethlehem Alloy and Special Steels." Bethlehem, Pa., The Author, 1935, 375 pages.
7. COLBECK, E. W. and D. HANSON: The Hardening of Silico-Manganese Steels, *J. Iron Steel Inst.*, v. 109, 1924, pp. 377-407.
8. MATHEWS, J. A.: Discussion. *Ibid.*, v. 109, 1924, pp. 404-405.
9. GILL, J. P.: "Tool Steels." Cleveland, Am. Soc. Metals, 1934, 136 pages.
10. PALMER, F. R.: "Tool Steel Simplified." Reading, Pa., The Carpenter Steel Co., 1937, 316 pages.
11. GILLET, H. W., and E. L. MACK: "Molybdenum, Cerium, and Related Alloy Steels." New York, Chemical Catalog Co., 1925, 295 pages.
12. HARRISON, R.: The Influence of Silicon on Nickel Steel, *J. Iron Steel Inst.*, v. 124, 1931, pp. 261-282.
13. SCHULZ, E. H., and F. BONSMANN: Über die Eigenschaften von silizium-legiertem Stahl in Form von Stahlguss (The Properties of Silicon Alloyed Steel in the Form of Steel Castings), *Stahl u. Eisen*, v. 50, 1930, pp. 161-168.
14. KÖRBER, F., and A. POMP: Mechanische Eigenschaften von Stahlguss bei erhöhten Temperaturen (Mechanical Properties of Steel Castings at High Temperatures), *Mitt. Kaiser Wilhelm Inst. Eisenf.*, v. 10, 1928, pp. 91-103.
15. RYS, A.: Legierter Stahlguss in Theorie und Praxis (Alloy Steel Castings in Theory and Practice), *Stahl u. Eisen*, v. 50, 1930, pp. 423-438.
16. FINLAYSON, A.: Low Alloy Steel Castings in Logging and Road Construction, *Metals & Alloys*, v. 8, 1937, pp. 239-244.
17. KINZEL, A. B., and R. W. MOORE: Graphite in Low-carbon Steel, *Metals Technology*, v. 1, September, 1934, 6 p. Also: *Am. Inst. Min. Met. Eng.*, Tech. Publ. 565. See also, WELLS, C.: Graphitization in High Purity Iron-Carbon Alloys, *Trans. Am. Soc. Metals*, v. 26, 1938, pp. 289-344, and AUSTIN, C. R., and B. S. NORRIS: Temperature-gradient Studies on Tempering Reactions of Quenched High-Carbon Steels, *Metals Tech.*, v. 5, 1938, A.I.M.E. Tech. Publ. 923, 23 pages.
18. SCHWARTZ, H. A.: Ferrous Alloy. U.S. Patent 2,069,423, February 2, 1937.
19. EDMUNDS, R. A.: Mixture for the Production of Malleable Iron. U.S. Patent 2,069,717, February 2, 1937.
20. BONTÉ, F. R.: Die Costs Cut with Graphitic Steels, *Am. Machinist*, v. 81, 1937, pp. 207-208.  
STUMPF, G. A., and F. R. BONTÉ: Graphitic Steels, *Industrial Heating*, v. 4, 1937, pp. 728-833.  
BONTÉ, F. R., and M. FLEISCHMANN: Developments in Graphitic Steels for Tools and Dies, *Metal Progress*, v. 31, April, 1937, pp. 409-413.
21. BONTÉ, F. R.: Ferrous Alloys and Methods of Manufacture. U.S. Patent 2,087,764, July 20, 1937.

## CHAPTER 5

### COPPER STEELS

Though not generally thought of as one of the important alloying elements, Cu is probably added to a larger tonnage of steel than any other alloying element. "Copper-bearing" steel, with about 0.20% Cu, is produced in huge tonnages for use where atmospheric corrosion is to be combatted; but, since such steel is not heat treated and much of it is rimmed steel, it is seldom classed with the quality alloy steels. There are no Cu steels in the S.A.E. series.

**Yield Strength Conferred by Copper.** Copper has rather recently gained greater importance owing to its inclusion, as a strengthening

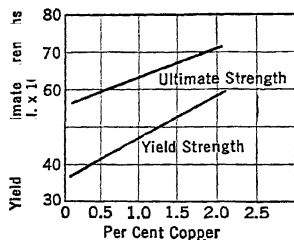


FIG. 1.—Effect of Cu in normalized low-C steels. (Epstein, Nead and Halley,<sup>1</sup> after Breiul.)

element, in several of the low-alloy, high-yield-strength, non-heat-treated steels. Its value in this regard, especially in producing high yield strength and high yield ratio will be discussed in Chap. 11, and has been brought out by Epstein, Nead, and Halley,<sup>1</sup> who show Fig. 1.

**Corrosion Resistance.** In these it serves also to give enhanced atmospheric corrosion resistance. With higher Cu content, ranging from 0.40 to 1.00%, and especially in conjunction with P, the corrosion resistance is improved over that of ordinary Cu-bearing steel. Although there is occasional application of high-Cu steels or Cu-P steels for submerged corrosion resistance, on the whole they are but little if any more resistant to such corrosion than analogous C steels. Only when the conditions of exposure are such that the rust film dries down into a sort of protective coating—which it does in atmospheric exposure—do the Cu steels show superiority. They *start* to rust just as fast as a C steel, even in atmospheric exposure.<sup>2</sup>

**Cost.** According to the market, the cost of Cu is either along with, or slightly above, that of Mn or Si, and at any rate more or less in the price class of Cr, so that there is reason to utilize it where it will serve, instead of more expensive alloying elements.

**Precipitation Hardening.** It has a very specific action in conferring precipitation-hardening possibilities as has been mentioned in Chaps. 8-10, Vol. I. The possibility of precipitation hardening is present when Cu is over 0.75%, 1.00 to 1.50% being about the best amount, and is most marked in the lower C steels (but up to 0.55% C will respond) where it can be utilized to raise the tensile and yield strengths as much as 20,000 lb./in.<sup>2</sup> without serious decrease in ductility. This hardening takes place throughout even heavy sections because of the very slow rate of cooling at which Cu may be retained in solid solution to be precipitated on holding at or reheating to the precipitation range of temperatures.

**Copper a Ferrite Strengthener.** Copper is not at all a carbide former, but up to about 0.75% under any conditions, and in larger amounts even on moderately slow cooling, it dissolves in the ferrite. Thus it may be employed alone or in conjunction with other elements as a ferrite strengthener. Copper exerts a mild graphitizing action, far less marked than Si and more of the order of, though not as strong as, that exerted by Ni.

**Retention on Re-melting.** Copper is less readily oxidized than Fe; hence all the Cu in scrap charged for steel making is recovered, just as is the case with Ni. Hence, as use of Cu-bearing steel increases and scrap from it returns, the Cu content tends to build up. The complete recovery favorably affects costs.

**Surface Cracking.** With large amounts, well over 0.50%, of Cu there is a tendency, in hot-rolled or forged products worked at a high temperature, toward edge cracking or surface checking, which can be overcome by very careful control of hot-working temperature, or the introduction of a little nickel.<sup>3</sup>

**Comparison of Copper and Nickel.** Bennek<sup>4</sup> compared wrought eutectoid steels without Ni or Cu, and with 0.25% of Ni or Cu, as to depth hardening on water quenching from 1510° of specimens about 1-in. diam. The plain C steel hardened to a depth of  $\frac{1}{8}$  in., while each of the steels with 0.25% Ni or Cu hardened to the same, greater, depth,  $\frac{1}{4}$  in. This would indicate that either element has the same effect on depth hardening, but information on the grain size or coarsening properties of the three steels, from which the hardenability might be deduced, was not included; so the proof is not complete. If this indication is correct, the Ni and Cu occurring as "residual metals"<sup>5</sup>

from remelting of Ni and Cu-bearing steel scrap would have a perceptible effect on hardenability.

The substitution of some Cu for Ni in wrought steels (by the use of Monel scrap instead of Ni) was suggested back in 1910 by Clamer<sup>6</sup> and such substitution, up to about 0.60%, has been made in many heat-treated steels produced at the Naval Gun Factory.<sup>7</sup> A larger substitution would bring in the possibility of specific action by Cu from precipitation hardening, but might be made without such effect if precipitation hardening is avoided. Finkl<sup>8</sup> suggests replacement of most or all of the Ni by Cu in a Ni-Cr-Mo steel for die blocks, claiming good depth hardening.

Harrison<sup>9</sup> has recently examined the effect of Cu in company with Cr. He finds that Cu may be useful in steels to be used in the hardened and tempered condition and comments particularly on the following comparison of Ni and of Cu as alloying elements in Cr steels, oil quenched from 1560° in  $1\frac{1}{2} \times \frac{3}{4}$  in., and drawn 2 hr. as shown:

No.	C	Si	Mn	Cr	Ni	Cu
22, Ni	0.30	0.28	0.44	0.90	3.50	
25, Cu	0.27	0.26	0.43	0.85	....	3.40

No.	Drawn at Deg.	Tensile	Yield	Elong. %	R. A. %	Izod Ft.-lb.	Brinell
22, Ni	930	157,000	138,000	19	57	21	321
25, Cu		177,500	153,000	17	49	16	353
22, Ni	1020	139,000	122,000	23	61	25	289
25, Cu		145,000	133,500	19	55	28	308
22, Ni	1110	127,000	105,500	25	66	59	264
25, Cu		131,000	119,000	21	57	61	281
22, Ni	1200	117,500	96,500	26	66	88	246
25, Cu		118,500	107,000	23	63	74	255

Harrison concludes that at least a part of the Ni in medium-C steels for quenching and tempering can be replaced by an equal weight of Cu without detriment. See further data from Harrison in Chap. 12.

**Effects upon Heat Treatment.** Below about 0.75% Cu, where precipitation hardening becomes possible, the presence of Cu does not affect the heat treatment of C or alloy steels to which it is added, except that the quenching temperature is lowered a bit since  $Ac_3$  is somewhat depressed, its effect in this respect being analogous to that of Ni. Copper slows down the transformation of austenite and thus acts to increase depth hardening. In amounts below those that allow precipitation hardening, its depth-hardening effect, though detectable, is not very strong.

Copper-C steels, water or oil quenched from about 1650° and then tempered above 800°, act normally save for slightly higher tensile and

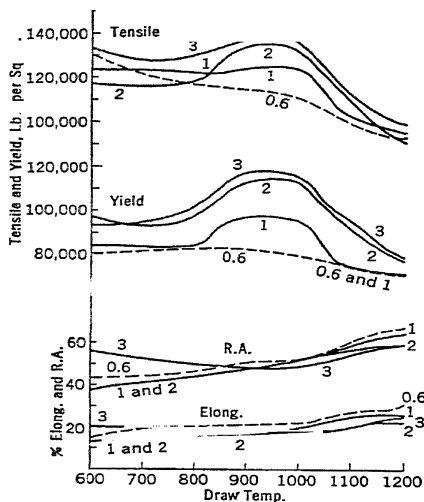


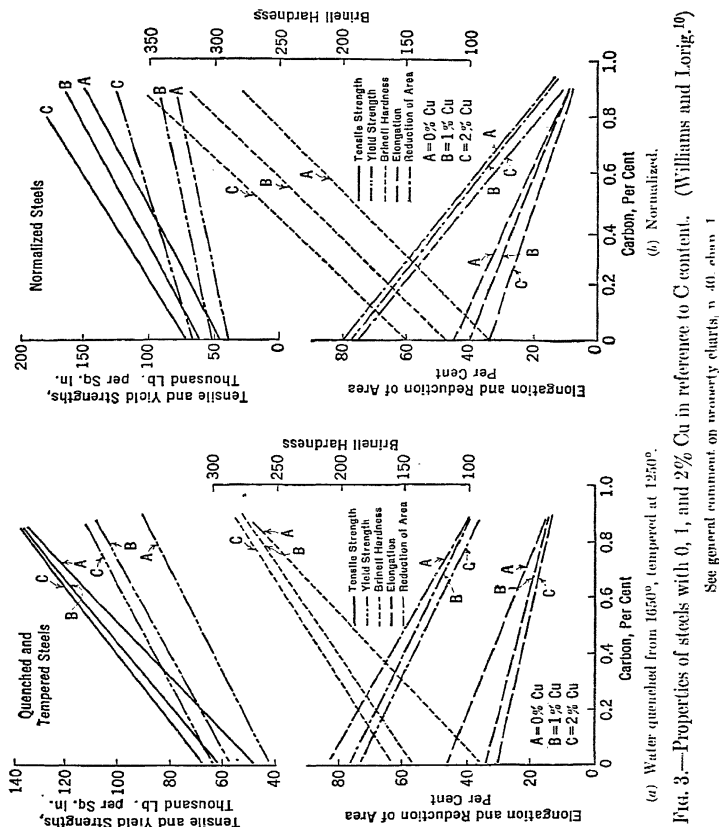
FIG. 2.—Properties of Cu steels, water quenched from 1500° and drawn as shown. (After Bird, Bethlehem Steel Co.)

	C %	Cu %
0.6.....	0.34	0.57
1.....	0.25	1.05
2.....	0.34	2.10
3.....	0.34	3.11

See general comment on property charts, p. 40, chap. 1.

yield strength and very slightly decreased ductility if the Cu is below 0.75%. If it rises to 1.00–1.50%, then the precipitation-hardening effect of Cu is superimposed on the C tempering of the hardened steel at 750–1100°, and brings out higher strength with somewhat lower ductility than would occur without the Cu. The maximum precipitation-hardening effect comes at about 900–1100° (time is a factor also), at which range the ductility shows a minimum. Above that range of temperature, re-solution of Cu starts (the steel becomes over-aged), and the softening effect of that action as well as that of C tempering are combined. These facts are brought out in curves based on early work by Bird of the Bethlehem Steel Co. (Fig. 2.) If

the tempering is at temperatures above the precipitation-hardening range, only the ferrite-strengthening effect of Cu is shown in quenched and tempered steels, as Fig. 3a shows. Similarly, in normalized steels, not reheated to develop precipitation hardening, only the ferrite-strengthening effect appears. (See Fig. 3b.)



(a) Water quenched from 1650°, tempered at 1250°.

(b) Normalized.

Fig. 3.—Properties of steels with 0, 1, and 2% Cu in reference to C content. (Williams and Lorig.<sup>10</sup>)

See general comment on accuracy charts, p. 40 above.

Inasmuch as the precipitation-hardening action is effective in very large sections, the possibility of adding to the core hardness and strength of shallow hardening steels is particularly interesting.

The effect of normalizing, with and without precipitation hardening, on a 0.40% C steel of various Cu contents is shown in Fig. 4<sup>10</sup> and



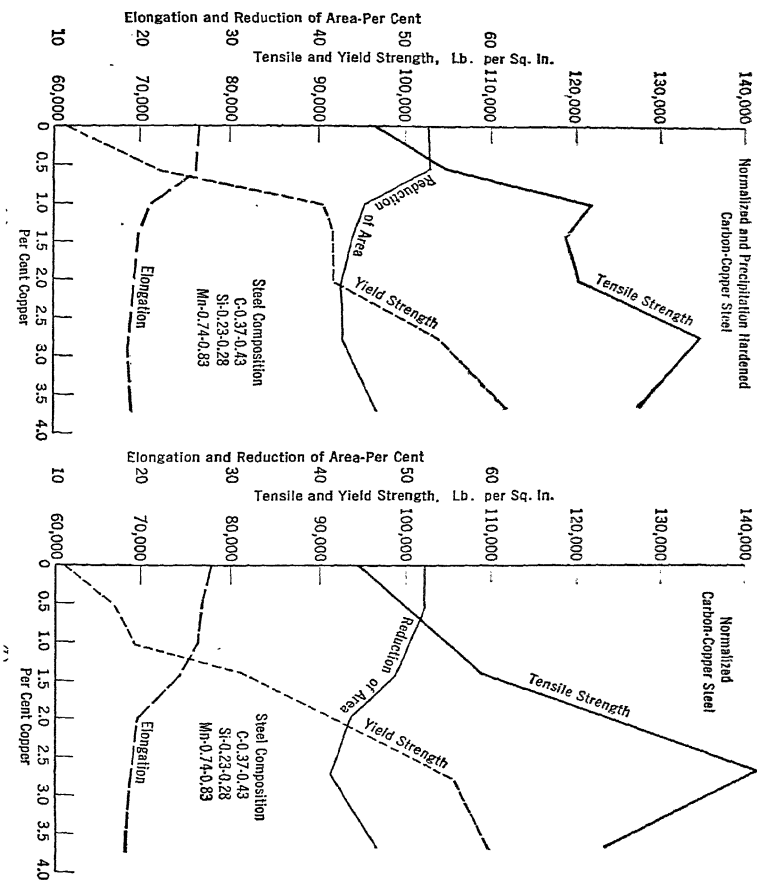


Fig. 4. Effect of Cu on properties of 0.43% C steels. (Williams and Lorig.<sup>10</sup>)

(a) Normalized, but not precipitation hardened.

(b) Normalized and precipitation hardened.

See general comment on property charts, p. 40, chap. I.

that on a normalized 0.56 C, 1.03% Cu steel is shown in Fig. 5. Figure 3b showed the effect of Cu and C in normalized steels and

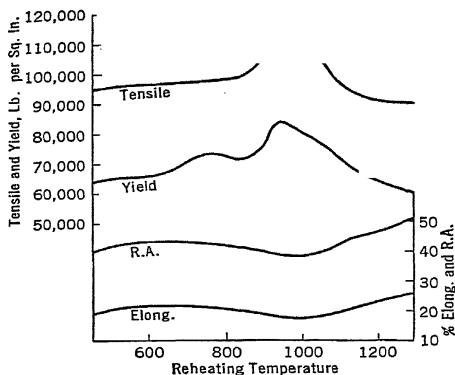


FIG. 5.—Effect of reheating for 4 hr. at temperatures shown, on a steel of 0.56 C, 1.03% Cu, normalized from 1490°. Battelle, from work for Copper and Brass Research Assn.)

See general comment on property charts, p. 40, chap. I.

Fig. 6 shows how the properties are changed when the optimum precipitation hardening or aging temperature is used.

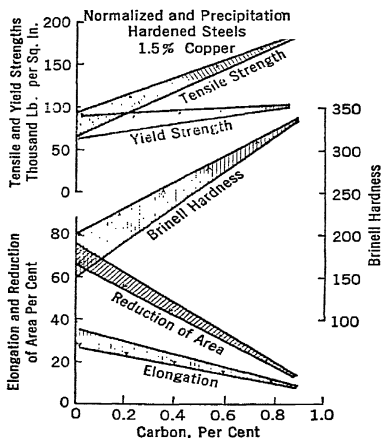


FIG. 6.—Magnitude and direction of changes in properties produced by precipitation hardening in 1.5% Cu steels of varying C contents. (Williams and Lorig.<sup>10</sup>)

See general comment on property charts, p. 40, chap. I.

**Copper in Case Hardening.** The ferrite-strengthening effect of Cu may be utilized in the core of case-hardening steels, since Epstein and Lorig<sup>11</sup> have shown that the presence of up to 3% Cu in 0.10–0.30% C steels does not decrease the depth or hardness of the case if the parts are free from a Cu layer formed as a result of oxidation. A

true Cu layer on the surface acts as a barrier to carburization, but Cu in the steel does not. Of course, reheating to the precipitation-hardening temperature of 800–1000° would soften the carburized case, though there might be uses for material with the gradation of hardness from case to core that would thus be produced.

**Copper Cast Steels.** Kinnear,<sup>12</sup> Finlayson,<sup>13</sup> Zeuge,\* and Lorig† have produced commercially or experimentally cast steels varying in C, Si, Mn, and Cu content. Finlayson favors low C, an example being:

0.15 C, 1.10 Mn, 1.10 Si, 1.90% Cu—annealed this gave 89,700,  
69,500, 29.5%, 57%  
0.14 C, 1.60 Mn, 1.10 Si, 1.80% Cu—annealed this gave 90,000,  
68,000, 26.5%, 51%

After precipitation hardening at 975° the latter gave 92,000, 77,500, 29.5%, 57%. These steels are extraordinarily fluid, a favorable matter from the foundry point of view.

Zeuge gives for steels normalized from 1650° precipitation hardened at 1000°:

0.27 C, 0.73 Mn, 0.34 Si, 0.97% Cu—106,000, 76,000, 21.5%, 43.5%  
0.32 C, 1.26 Mn, 0.46 Si, 1.92% Cu—137,000, 105,000, 16%, 39%

These are interesting when high yield strength is sought. Hamilton and Sims<sup>14</sup> also show the effect of precipitation hardening in Mn-Cu cast steel.

When precipitation hardening is not to be employed and the steel is designed to meet the requirements of 90,000 tensile, 60,000 yield, 22% elong., 45% R. A., 30 Izod, it is found that with normal Mn and C the combination of strength and ductility cannot readily be met by addition of Cu alone, the impact in particular failing to meet the requirements when the tensile specification is met.

According to "Vanadium Steels and Irons"<sup>15</sup> the addition of 0.10% V will turn the trick, since 0.29 C, 1.10–1.35 Mn, 1.05 Cu, plus 0.10% V, double normalized and drawn, averaged:

	TENSILE	YIELD	· ELONG. %	R.A. %	
750° draw.....	95,500	71,500	29	56	61
950° (precipitation hardened).....	107,000	86,000	25	50	31

The properties can be readily enough met with sufficient Ni, and V, or Mo, as has been brought out in Chap. 3. But the Cu steels would have an advantage in cost over the Ni, Mo and MnV used for this

\*Cited in 21.

† Unpublished Work. Battelle Memorial Institute.

grade of steel. A so-called Mn-Cu steel (reported by Armstrong<sup>16</sup>) misses the tensile requirement only by a hair, exceeds the ductility requirement, and has a fine impact. With a trace more C it might meet the specification even without V. Armstrong's comparison of such a steel with various Ni steels is shown in the following table. The last steel on the table is reported by Duma<sup>17</sup> and was double normalized from 1650° and 1525° and drawn at 1100°. It seems to be the same as Armstrong's No. 12, raising the question whether Armstrong's comparison is for a low Ni or a Ni-free steel.

STEELS NORMALIZED TWICE FROM 1650°, DRAWN 1200°

No.	% C	% Mn	% Si	% Ni	% Cu	% Mo	% V	Tensile	Yield	Elong. %	R.A. %	Izod.
12	0.24	1.02	0.31	.....	1.26	.....	.....	87,250	72,750	25	57	62
18	0.23	1.44	0.19	1.12	.....	.....	.....	86,200	70,000	29	58.5	61
15	0.22	1.16	0.16	2.32	.....	.....	.....	89,500	71,000	28	54	52
22	0.30	0.94	0.28	1.45	.....	.....	0.17	95,500	70,500	26	54	45
25	0.23	1.09	0.24	2.21	.....	0.33	0.13	96,000	76,500	24	58	69
11	0.27	0.99	0.34	1.26	1.17	.....	.....	95,000	73,250	25	52	43
9	0.24	1.02	0.31	0.64	1.26	.....	.....	87,250	72,750	25	57	29 to 47

The possibility of substituting some of the Ni by Cu, in a Ni-Mn steel of this class, is evident.

Avoidance of the expensive elements employed to secure the desired properties in the steels mentioned above may, according to unpublished work by Lorig at Battelle Memorial Institute, be brought about by the substitution of Cr for some of the Mn in these Cu steels, coupled with grain-size control for ductility by the use of Ti or Al as shown in the table below for steels double normalized, 1650-1575°, or double normalized and drawn at 1250°.

Treatment	% C	% Mn	% Si	% Cu	% Cr	% Ti	% Al	Tensile	Yield	Elong. %	R.A. %	Izod.
N. N.	0.26	0.80	0.40	1.30	0.50	0.10	.....	96,500	70,000	26	57	45
N. N. D.	.....	.....	.....	.....	.....	.....	.....	88,000	63,500	28	62	57
N. N.	0.32	0.80	0.40	1.20	0.50	0.10	.....	105,000	73,000	24	54	39
N. N. D.	.....	.....	.....	.....	.....	.....	.....	96,000	63,000	24	53	45
N. N.	0.24	0.80	0.40	1.30	0.55	.....	0.10	95,000	70,000	26	54	54
N. N. D.	.....	.....	.....	.....	.....	.....	.....	87,500	63,000	27	57	58
N. N.	0.33	0.88	0.40	1.25	0.54	.....	0.10	104,500	71,500	22	48	41
N. N. D.	.....	.....	.....	.....	.....	.....	.....	94,000	64,000	24	51	44
N.N.D.*	0.28	1.05	0.40	1.00	.....	0.20	.....	93,000	70,000	28	52	43
N.N.D.*	0.28	1.10	0.40	1.00	.....	0.20†	.....	96,500	73,500	27	52.5	40

\* 1700°, 1600°-750°, 1 hr.

† Amount added.

**Manganese-Copper-Titanium Cast Steels.** From the discussion on Mn-Ti cast steel, Chap. 3, it will be appreciated that a lower Ti content will probably suffice and that single normalizing with or without a draw (since the draw given was too low to affect the properties) may be satisfactory for the last steel in the table. The C content of that last steel is probably as high as is desirable.

To check some of these conclusions the following were single normalized from 1600° with no draw.

C	Mn	Si	Cu	Ti	Tensile	Yield	Elong. %	R.A. %	Izod
0.23	1.13	0.29	0.99	*	91,000	68,000	28.5	57.5	45
0.28	1.09	0.35	1.02	*	98,000	71,500	26	49	36
0.27	1.34	0.38	1.20	*	104,000	79,000	25.5	54	29
0.29	1.17	0.37	1.19	*	100,500	75,500	26	52	29
0.31	1.12	0.33	1.00	†	97,000	69,000	25	47.5	33

\* 0.20% Ti added.

† 0.05% Ti added.

Comparing these figures with those for Mn-Ti in Chap. 3, it appears that dropping the C and Mn and adding Cu may be advantageous in improving yield strength at a given ductility as well as in improving the steel from the foundry point of view. Experimentation with the balancing of C, Mn, Si, and Cu, coupled with the use of Ti, or some other grain-refining element, should produce a steel of good behavior in the foundry and with excellent mechanical properties, at low cost.

A cast steel of about 1.15 Mn, 1.00 Cu, 0.35% Mo is used for water quenching. Care should be taken to remove the castings from the water while they are still warm and to transfer them at once to the draw furnace to avoid cracking. By varying the draw temperature a wide range of properties may be secured.

**Ford Crankshaft "Steel" (Cu-Si-Cr).** A "steel" that has attracted much attention is the alloy used, at present writing, in the Ford V8 crankshaft. One has some hesitation in calling it steel, since it is not malleable as cast but is cast white, graphitized like pearlitic malleable Fe, and then heat treated like steel. After malleabilization it falls into a somewhat similar category to graphitic steel but lacks the forgeability of that material. The crankshaft alloy could have been discussed under Si steel, but since Cu is the chief alloying element it is best dealt with here.

The alloy contains about 1.35–1.60 C, 0.60–0.80 Mn, 0.85–1.10 Si, 1.50–2.00 Cu, and 0.40–0.50% Cr. After being held at 1650° for 20

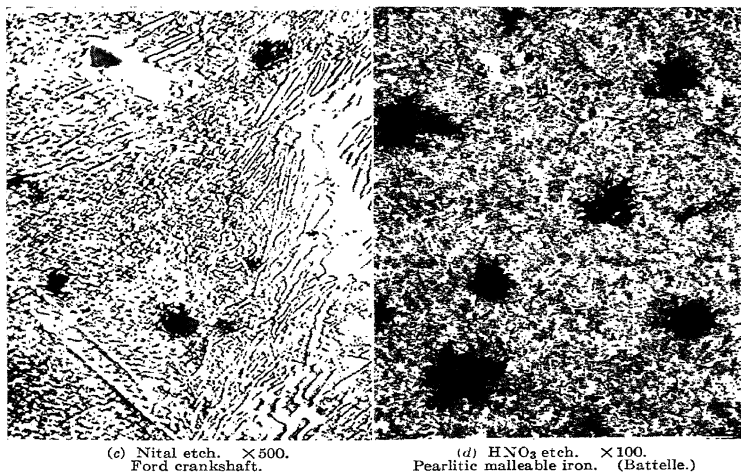
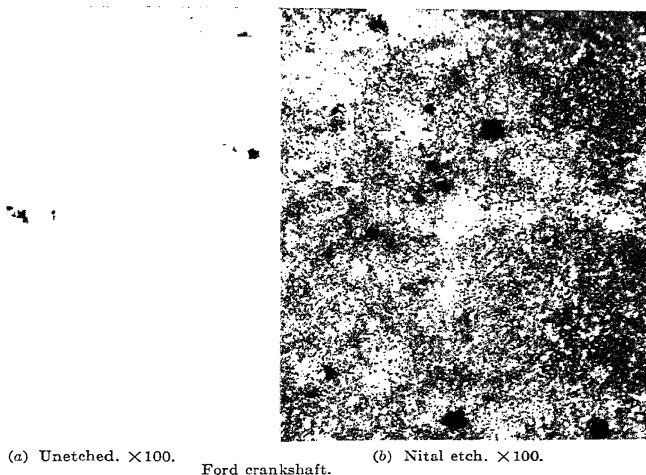


FIG. 7.—Structure of Ford crankshaft compared with that of malleable iron. The commercial crankshaft contained, total C, 1.62; Mn, 0.74; Si, 0.99; Cu, 1.62; Cr, 0.51%. The malleable iron contained, total C, 1.89; combined C, 0.66; Si, 1.00; Cu, 0.95%, and was held 1 hr. at  $1550^\circ$ , oil quenched, and drawn 2 hr. at  $1300^\circ$  to 229 Brinell.

Note the smaller nodules in the Ford material. Its Cr content confers resistance to graphitization.

min. the castings are normalized by air cooling till they are below the critical (i.e., at about  $1200^{\circ}$ ), reheated to  $1480^{\circ}$ , held one hour, cooled in the furnace for another hour, and removed at about  $1000^{\circ}$ . The structure is shown in Fig. 7*a, b, c*. For comparison, Fig. 7*d* shows the structure of a heat-treated, pearlitic Cu malleable iron, with a C content above the Ford range; Fig. 8 shows a fully malleabilized malleable iron.

In test bars Cone<sup>18</sup> and McCarroll<sup>19</sup> report that the Ford alloy shows about 107,500 tensile, 92,000 yield, 1.75% elong., 2.5% R. A., 269 Brinell. Specimens cut from a production crankshaft of 1.66 C, 0.89 Si, 0.70 Mn, 1.78 Cu, 0.65% Cr tested by Russell<sup>20</sup> gave 78,000

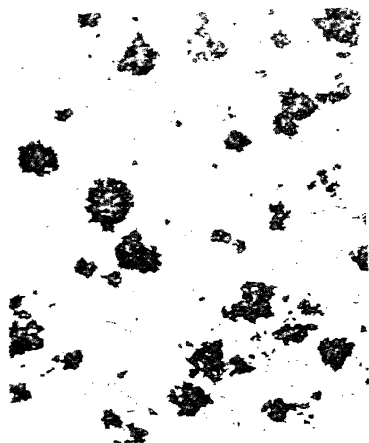


Fig. 8.—Malleable iron.  $\times 100$   
(Battelle.)

Fully malleabilized iron of C, 1.86; Mn, 0.59, Cu, 1.04; Si, 1.04%. The matrix is ferrite; whereas that of Figs. 7*b* and 7*d* is pearlite.

tensile with no elongation or reduction of area, but had an endurance limit of 44,000 in the usual specimen and 31,000 on a specimen with a square notch. The latter figure was slightly improved by giving the shaft a precipitation-hardening treatment. The low notch sensitivity of the alloy is especially noteworthy.

A similar alloy with slightly lower Mn and no Cr is used for brake drums, the heat treatment being  $1600^{\circ}$  for 30 min., cool rapidly (in furnace) to  $1450^{\circ}$ , cool slowly (2 hr.) to  $1350^{\circ}$ , then to  $1000^{\circ}$  in 1 hr. This gives 85,000 tensile, 70,000 yield, 6% elong., 195–230 Brinell.

In these alloys the Cu is used to give fluidity in casting, to aid graphitization, and, of course, to aid in securing yield strength.

**Summary.** Copper is a ferrite-strengthening element, favoring high yield strength in as-rolled low-C steels. It does not enter the

carbide, and has only a mild graphitizing action. It does not form a gamma loop. It lowers  $Ac_3$  to about the same extent that Ni does. It can apparently be substituted for up to around 0.75% of Ni, with almost equivalent results, though this ability has not been widely utilized. The effect upon depth hardening is not clearly defined, but may be of the same order as that of equivalent small amounts of Ni, i.e., not sufficient to be very effective unless accompanied by other alloys.

Its main uses are for atmospheric corrosion resistance and to confer high yield strength in low-C, non-heat-treated steels.

Besides its specific influence on atmospheric corrosion resistance, it has, in amounts of say 0.75–2.00%, a very specific effect in supplying ability to precipitation harden, and this is applicable to very large sections.

It is useful in cast steels.

It also finds use in certain high-C, high-Si “steels” that are cast as white iron and then graphitized.

The Cu steels are discussed in more detail in “The Alloys of Iron and Copper,”<sup>21</sup> and a concise discussion is given in “Copper in Cast Steel and Iron.”<sup>22</sup>

## BIBLIOGRAPHY

1. EPSTEIN, S., J. H. NEAD, and J. W. HALLEY: Choosing a Composition for Low-Alloy High-Strength Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 120, 1936, pp. 309–345. Also: *Metals Technology*, v. 3, 1936, Tech. Publ. 697, 31 p.
2. BRITTON, S. C.: Some Properties of Commercial Steel Sheets Containing Additions of Copper, Manganese, Chromium, and Phosphorus, *J. Iron Steel Inst.*, v. 135, 1937, pp. 161–185.
3. LORIG, C. H.: Discussion, *Trans. Am. Inst. Min. Met. Eng.*, v. 105, 1933, pp. 165–167.
4. BENNEK, H.: Einfluss kleinster Beimengungen von Kupfer und Nickel auf Kohlenstoffstähle (Influence of Small Quantities of Copper and Nickel on Carbon Steels), *Tech. Mitt. Krupp*, v. 2, 1934, pp. 129–133.
5. SULLIVAN, J. D., and R. A. WITSCHY: Residual Metals in Open-hearth Steels, *Metals & Alloys*, v. 8, 1937, pp. 99–100. See also preceding volumes of this journal for earlier references.
6. CLAMER, G. H.: Cupro-Nickel Steel, *Proc. Am. Soc. Test. Mat.*, v. 10, 1910, pp. 267–279.
7. GILLET, H. W., and E. L. MACK: “Molybdenum, Cerium, and Related Alloy Steels.” New York, Chemical Catalog Co., 1925, 295 pages.
8. FINKL, W. F.: Die Block, U.S. Patent 2,104,979, January 11, 1938.  
FINKL, W. F.: Steel Alloy, U.S. Patent 2,104,980, January 11, 1938.
9. HARRISON, R.: The Effect of Copper on Some Alloy Steels, *Iron Steel Inst.* Advance Copy 4, May, 1938, 38 pages.



10. WILLIAMS, C. E., and C. H. LORIG: The Role of Copper in Wrought Steel, *Metals & Alloys*, v. 7, 1936, pp. 57-63.
11. EPSTEIN, S., and C. H. LORIG: Note on the Carburizing of Copper Steel, *Ibid.*, v. 6, 1935, pp. 91-92.
12. KINNEAR, H. B.: Alloy Steel and Method of Producing the Same, U.S. Patent 1,607,056, November 16, 1936.
13. FINLAYSON, A.: Low Alloy Steel Castings in Logging and Road Construction, *Metals & Alloys*, v. 8, 1937, pp. 239-244.
14. HAMILTON, W. C., and C. E. SIMS: Steel, U.S. Patent 1,742,857, January 7, 1930.
15. Vanadium Corporation of America: "Vanadium Steels and Irons." New York, The Author, 1937, 189 pages.
16. ARMSTRONG, T. N.: Properties of Some Cast Alloy Steels, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 286-318.
17. DUMA, J. A.: Effect of Titanium on Some Cast Ferrous and Nonferrous Metals, *Ibid.*, v. 25, 1937, pp. 788-825.
18. CONE, E. F.: The Story of the Ford Cast Crankshaft, *Metals & Alloys*, v. 6, 1935, pp. 259-263.
19. Comment by R. H. MCCARROLL, Ford Motor Company, *Ibid.*, v. 7, 1936, pp. 323-324.
20. RUSSELL, H. W.: Resistance to Damage by Overstress of Precipitation-Hardened Copper-Steel and Copper-Malleable, *Ibid.*, v. 7, 1936, pp. 321-323.
21. GREGG, J. L., and B. N. DANILOFF: "The Alloys of Iron and Copper." New York, McGraw-Hill Book Co., 1934, 454 pages.
22. Copper Development Association: "Copper in Cast Steel and Iron." London, The Association, 1937, 136 pages (C.D.A. Publication 29).
23. ALEXANDER, M.: Copper-Steel Castings, in Third Report of the Steel Castings Research Committee, Iron Steel Inst., Oct., 1938.

## CHAPTER 6

### THE PLAIN NICKEL STEELS

The plain Ni steels formed the pioneer class of alloy steels, developed before heat treatment became common, and were utilized to give, in the as-rolled condition, higher strength and greater toughness than could be secured from plain C steels. For example, a steel of 0.25 C, 3.50% Ni has equal strength, higher yield, and greater ductility compared with 0.45% C steel.

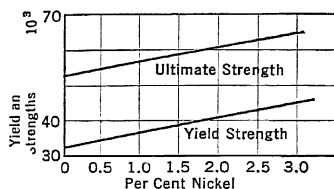


FIG. 1.—Effect of Ni in normalized low-C steels. (Epstein, Nead and Halley.<sup>16</sup>)

The word most used in describing Ni steels is “tough.” This toughness is due to the ferrite-strengthening power exerted by rather large amounts of Ni. The Ni enters the ferrite only; it does not form carbides in steel.

The increase in strength produced by adding Ni to carbonless Fe is relatively low compared to that conferred by other elements previously discussed, as Fig. 1 shows.

Nickel is less oxidizable than Fe, so no Ni is lost in melting. All Ni in scrap is recovered, and the steel-making process is not complicated by its presence.

**Normalized Steels.** Normalized steels, in small sizes, show the relations schematically indicated in Fig. 2. The lower boundary of the bands is for plain C, the upper for 5% Ni steel. As the Ni increases from zero to 5%, the tensile, yield, and reduction of area increase toward the top of the band; elongation shows less change. The impact resistance is very materially increased by the increase in ductility for a given hardness. The outstanding characteristic of Ni-strengthened ferrite is its toughness. This toughness is retained to very low temperatures instead of dropping off sharply at some moderately low temperature, as will be discussed in Chap. 19.

**S.A.E. Plain Nickel Steels.** The S.A.E. plain Ni steels show a range of Ni in the carburizing grades but hold to 3.5% Ni in the structural grades, with slightly higher Mn in the higher carbons.

No.	% C	% Mn	% Ni
2015	0.10-0.20	0.30-0.60	0.40-0.60
2115	"	"	1.25-1.75
2315	"	"	3.25-3.75
2515	"	"	4.75-5.25
2320	0.15-0.25	"	3.25-3.75
2330	0.25-0.35	0.50-0.80	"
2335	0.30-0.40	"	"
2340	0.35-0.45	0.60-0.90	"
2345	0.40-0.50	"	"
2350	0.45-0.55	"	"

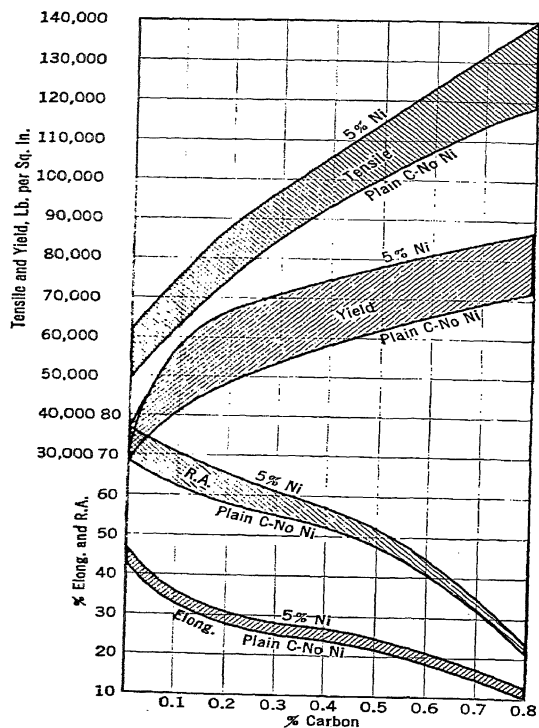


Fig. 2.—Effect of addition of Ni, in normalized steel, small section.  
See general comment on property charts, p. 40, chap. 1.

**Low-Carbon, Nickel Steels.** Rather low-C, Ni steels are used, as-rolled, for the strength and toughness conferred by Ni, e.g., 0.20 C, 2.25% Ni steel is used for locomotive boilers. In 1-in. plate this gives 76,000, 46,000, 27, 55; 0.15 C, 1.50% Ni rod for bolts, in 1-in. rods, shows 70,000, 45,000 to 50,000, 30, 65. On water quenching and tempering, the latter steel rises to 80,000, 64,000, 30, 70. Even so low a C as 0.05% with 2% Ni, a steel used for stay bolts, shows some response to quenching. In the hot-rolled condition 48,000, 35,000, 30, 65, and water quenched and drawn 58,000, 45,000, 22, 68 are reported for such steels.

The ferrite-strengthening effect of Ni may be noted in the increase produced in endurance ratio. The low-C, Ni staybolt steel is reported as giving 55,000 tensile, 35,000 yield, 32% elong., 62% R.A. and 40,000 endurance limit in rotating bending tests. That is, in repeated rotary bending the steel may be stressed above its yield strength without failure.

**Medium-Carbon, Normalized Nickel Steels.** Heavy forgings, as for railway axles, are made with slightly higher C and Ni, for example, 0.25 C, 0.85 Mn, 2.65% Ni, and are normalized from around 1475°, drawn at 1050–1200°. Results cited for specimens from the middle of a 10-in. section are 87,000, 62,000, 30, 62.

Maurer and Korschan<sup>1</sup> cite a 0.23 C, 0.25 Si, 0.50 Mn, 1.90% Ni steel in very large forgings. Transverse specimens from normalized forgings gave the following:

Diameter Forging	Tensile	Yield	Elong. %	R.A. %
36 inches	71,000	40,500	19.5	20
46 inches	73,000	42,000	18	26
57 inches	74,000	43,000	15.5	20

The amount of working was smaller as the forging diameter increased.

**Heat-Treated Nickel Steels.** In quenched and tempered Ni steels (Fig. 3) analogous improvement in toughness is observed over the C steels, as is found in the normalized steels. Besides this, Ni in sufficient amount confers marked depth-hardening properties, as is shown by the critical cooling rate comparison of Fig. 4. The tensile strength and reduction of area of 1045 (0.45% plain C) and 2340 (0.40 C, 3.50% Ni) compared in Fig. 5 (curves N<sub>1</sub> and C<sub>1</sub>) for the water-quenched, 800° tempered condition for both show better penetration of hardening as well as a higher strength level for equal or superior ductility of

the Ni steel, whereas curves  $N_2$  and  $C_2$  show that oil quenching of the Ni steel produces greater strength than water quenching of the plain C steel.

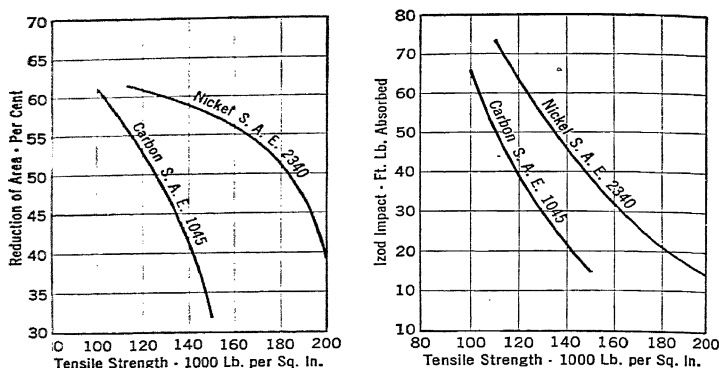


FIG. 3.—Ductility and impact resistance of medium C steel, water quenched, and of a similar steel with Ni, oil quenched, both drawn to the same tensile strength. (Nickel Alloy Steels.<sup>5</sup>)

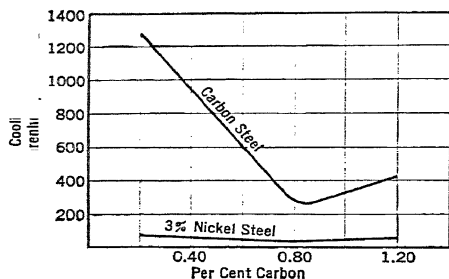


FIG. 4.—Effect of 3% Ni on critical cooling rate. (Nickel Alloy Steels.<sup>5</sup>)

With sufficiently high quenching temperature the critical rates may actually decrease, instead of increasing as here shown, at the higher C contents. Compare Fig. 9, chap. 6, Vol. I.

In ordinary commercial practice the cooling rates probably are as here shown.

**Nickel-Vanadium Steels.** Introduction of V may be made for grain-size control. Data for a Ni-V steel are in the table, p. 117.

If the Ni is raised too far in relation to the C, the steels become air hardening, i.e., are martensitic. Further increase makes the steel austenitic. This is shown in Fig. 6. According to the "Welding Handbook"<sup>2</sup> a steel of 0.18 C, 2.50% Ni is sufficiently weld hardening so that welded joints in such a steel require stress-relief annealing.

## NICKEL-VANADIUM STEEL

C 0.29%    Ni 3.41%    Mn 0.45%    Si 0.09%    V 0.20%

Treatment		Tensile	Yield	Elong. %	R.A. %
Annealed	1470°	107,500	73,000	23.5	55.5
	Oil-quenched				
1600°	1160°	148,500	126,000	18.0	58.0
	1110	150,000	128,500	17.5	57.5
	1020	151,500	132,500	16.0	57.0
	930	162,000	144,000	14.5	52.5
	840	178,000	157,000	13.0	52.5
	750	193,000	163,000	12.0	50.0
1520°	1160°	137,700	123,000	16.0	59.0
	1110	140,700	125,500	17.5	54.0
	1020	148,000	127,000	16.5	55.0
	930	155,000	135,000	15.5	57.0
	840	167,000	146,500	14.0	55.0
	750	181,000	163,000	14.0	53.5
Water-quenched					
1600°	1160°	148,000	126,500	18.5	58.0
	1110	154,000	133,000	15.0	59.0
	1020	156,500	136,500	14.0	54.5
	930	161,000	146,500	14.5	56.5
	840	186,500	173,500	13.0	52.5
	750	195,000	176,500	12.0	52.0
1520°	1160°	140,000	128,500	18.5	59.5
	1110	146,000	132,000	14.0	57.5
	1020	154,500	134,000	15.5	56.5
	930	160,500	144,500	15.0	51.5
	840	184,500	177,000	13.0	53.0
	750	199,500	182,500	12.0	50.0

These data are compared with 2330 in Fig. 4, Chap. 8. The higher quenching temperature, made possible by the use of V, is a factor in the production of greater strength.

The behavior of even low-C, Ni steels on mild quenching in heavy sections is shown by the following data for tests at midsection of railway driving axles 10.5-in. diam., 3.5-in. bore, with the steels normalized from 1475°, oil quenched from 1475°, drawn 1200–1300°

%C	%Mn	%Ni	TENSILE	YIELD	ELONG. %	R.A. %
0.23	0.86	2.75	92,000	70,000	27	58
0.34	0.92	2.79	104,500	73,000	26.5	51

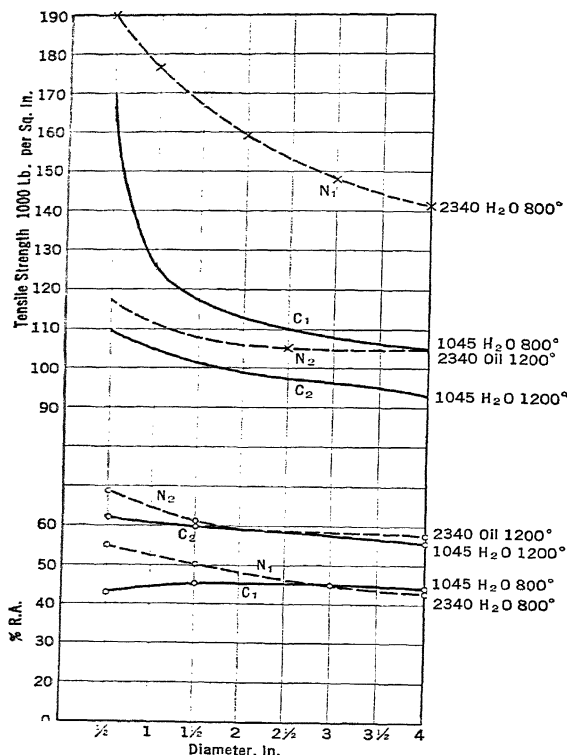


Fig. 5.—Mass effect in 0.45% C and 0.40 C, 3.50% Ni steels, quenched and drawn as shown. Longitudinal specimens cut midway between surface and axis. (Nickel Alloy Steels.<sup>5</sup>)

See general comment on property charts, p. 40, chap. 1.

The standard Ni content is 3 to 3.5%. However, much of the improvement due to Ni is produced with somewhat lower Ni. Guzzoni<sup>6</sup> compares steels water quenched from 1600°, drawn at 1110°, as follows:

%C	%Ni	TENSILE	YIELD	ELONG. %
0.21	3.0	122,000	95,000	16
0.27	1.8	106,500	78,000	14

<sup>6</sup> Gage length different from that used in U. S. making the elongation values appear low.

According to Müller-Hauff and Stein,<sup>4</sup> steels of 0.35 C, 0.55 Mn, and 2 and 3% Ni compare as follows:

	Tensile	Yield	Elong. %	R.A. %
Annealed 2%.....	85,000- 71,000	64,000-50,000	19-22 *	60-70
Annealed 3%.....	93,000- 78,000	71,000-57,000	18-23	60-70
Heat treated 2%...	100,000- 85,000	78,000-64,000	18-23	60-70
Heat treated 3%...	114,000-100,000	93,000-78,000	15-20	60-70
Heat treated 2%...	114,000-100,000	93,000-78,000	15-19	60-65
Heat treated 3%...	128,000-114,000	107,000-93,000	13-15	55-65

\* Gage length different from that used in U. S. making the elongation values appear low.

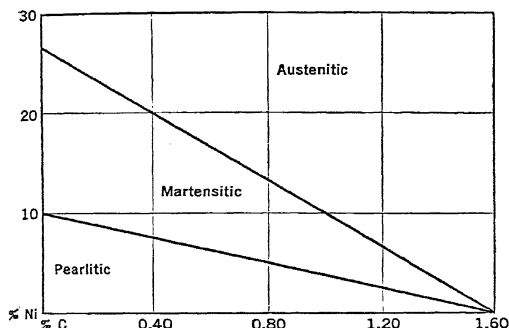


FIG. 6.—Influence of contents of Ni and C on the structure of Ni steel as cast.

The heat treatments are not given but are apparently the same for each pair. Five per cent Ni of the same C and Mn was also compared at the same treatments, the tensile and yield values cited being identical with those for 3% and the ductility being only a trace higher.

A 1 Ni, 0.40% C steel has been used, especially abroad, oil quenched and tempered, but in general a plain Ni constructional wrought steel to be used uncarburized but heat treated contains somewhere from 3 to 3.5% Ni. This amount appears to give the best balance between the desirable properties and the cost of Ni added.

Comparison of 0.40 C, 0.65 Mn, 1% Ni steel with 0.30 and 0.40 C, 3.50% Ni steels (2330 and 2340) on 1 $\frac{1}{8}$ -in. diam. bars, according to "Nickel Alloy Steels"<sup>5</sup> is made in the table, p. 120.

The effects of small amounts of Ni on the properties of any but the low-C carburizing steels and on the depth-hardening propensities are not well described, despite the long period of use of Ni as an alloying element. Lower Ni is generally used in steel castings and both



C, %	Ni, %	Medium	Quenching Temp.	Draw Temp.	Tensile	Yield	Elong. %	R.A. %	Izod
0.30	3.50	Water	1450°	800°	171,000	150,000	15	57	
				900°	150,000	130,000	17	60	
				1050°	125,000	105,000	22	63	
				1250°	100,000	80,000	25	65	
0.30	1.00	Water	1560°	600°	170,000	135,000	12	35	15
				900°	152,000	117,000	17	48	27
				1100°	125,000	90,000	23	60	48
				1300°	100,000	65,000	30	68	80
0.30	3.50	Oil	1450°	1000°	125,000	78,000	22	60	55
				1200°	105,000	70,000	25	64	85
0.40	3.50	Oil	1425°	1050°	125,000	110,000	22	62	55
				1250°	110,000	85,000	26	62	72
0.30	1.00	Oil	1560°	900°	122,000	90,000	22	56	26
				1200°	100,000	67,000	27	63	47

lower and higher Ni in carburizing steels. In wrought steels for heat treatment the lower Ni steels are usually reinforced by the addition of some other alloying element, as will be discussed in Chap. 12.

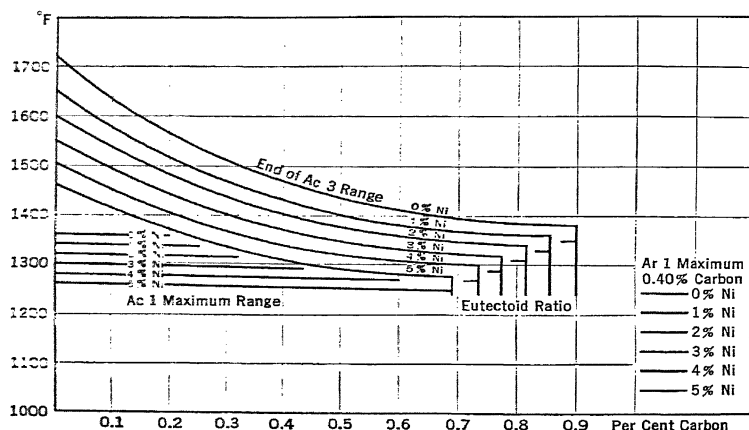


FIG. 7.—Critical range diagram for Ni steels.

(The eutectoid ratio shown for 0% Ni is higher than that accepted at present.)

**Effect of Nickel on Heat Treatment Temperatures.** An important point in the heating of the Ni steels for annealing, normalizing, or quenching is that  $Ac_3$  is materially reduced by Ni, the effect being shown in Fig. 7. This also shows the decrease in the eutectoid ratio as Ni increases. In pure Fe-Ni alloys Smith <sup>6</sup> reports the critical point on heating as 1425° for 2.0% Ni, 1315° for 3.13% Ni, and 1150° for 5.3% Ni.

Thus a steel containing Ni will be heated for quenching to a much lower temperature than an analogous C steel because of the position of  $Ac_3$ , just as the Si steels are heated higher because of the position of  $Ac_3$  in those steels.

**Grain Size.** This lower temperature of heating helps Ni steels to remain fine. Thus, while Ni steels are not in themselves inherently and of necessity fine grained, the conditions under which they are heat treated tend to avoid coarsening. Added to this is the fact that the Ni-containing austenite, other things being equal, seems more sluggish to coarsen than does plain C austenite. However, there are Ni steels of low-coarsening temperature and of high-coarsening temperature, corresponding to the deoxidation practice used in the melting, even though the general tendency is for the presence of Ni to decrease the propensity toward coarsening. The better inherent toughness of the Ni steels for a given grain size also makes it possible to heat them higher without bad results. Thus on the whole the Ni steels are less touchy to overheating than C steels. However, additions for grain-size control serve a useful purpose in these, as in other steels.

Moreover, Ni-bearing austenite is somewhat more sluggish in solution of cementite and in diffusion of C for homogenization than plain C austenite, so that the Ni steels are best heated further above their  $Ac_3$  points than is the case with C steels. The  $Ar_1$  point is strongly depressed, that for a 0.30 C 3.50% Ni steel being about 1100°. These various peculiarities of Ni steels make it possible to homogenize and quench a Ni steel with different heating methods. When a steel that has been finished cold in hot working, or that starts out with a coarse structure, is to be refined, (1) a low quenching temperature, but a long soaking time, may be used, or (2) a high quenching temperature with a normal soak, and, (3) when cracking is feared as a result of high quenching temperature, double treatment, normalizing, or mild quenching from the high temperature followed by a reheat and quench from the low one, or even (4) heating to such temperature and for such time as may be required, but cooling fairly slowly till a temperature a little above  $Ar$  is reached before quenching. Number 4 has much to commend it from the point of view of avoiding cracking in quench-

ing, while the double treatment of high-temperature normalizing followed by low-temperature heating for quenching, with the quench delayed to the lowest feasible temperature, tends to bring out the best properties. Compare Angell.<sup>7</sup>

**Carburizing Steels.** The lowered  $Ac_3$ , by which austenite is produced at a lower temperature than with C steels, allows carburizing at low temperatures if desired. By so doing the C content of the case can be held down and a tough case produced. Or, by using a steel high in Ni, such as 5%, only about 0.70% C is needed (as Fig. 7 shows) to produce an eutectoid case, and only about 0.80–1.00% (as Fig. 6 shows) to make the case martensitic even on slow cooling, without quenching. This is sometimes desired. With high enough Ni and C the case may be given some austenite along with the martensite, if the type of case toughness thus produced is desired and the lower hardness can be endured. The rate of carburization and depth of case are not greatly affected by presence of Ni though most authorities think they are slightly reduced. Grain coarsening of the core tends to be reduced by Ni, other things being equal.

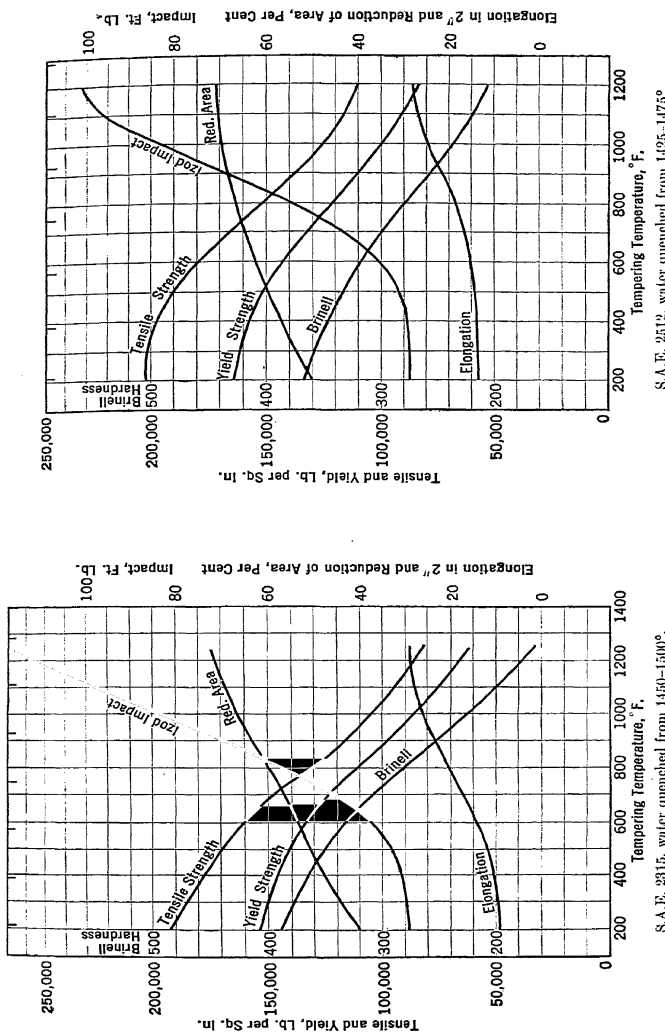
The lower Ni carburizing steels 2015, 2115, and 2315 have some of the advantages of the high Ni type, such as core toughness, ability to be carburized at low temperatures, etc., in ratio approximating to their Ni content. Since Ni is a ferrite strengthener only, the addition of carbide-forming elements to produce harder carbide is common. Such steels will be discussed in Chap. 12.

**Property Charts.** Property charts for plain Ni, heat-treated steels, as given in "Nickel Alloy Steels,"<sup>8</sup> are shown in Figs. 8–11.

**Replacement of Nickel.** Because of the cost of the Ni content of earlier steels there has been constant effort to use smaller amounts, replacing the omitted Ni by other elements, or substituting Ni-free steels. This tendency in England has been commented upon by Jones.<sup>9</sup> In Germany, recent pressure against the use of imported materials has accelerated the effort for substitution.<sup>9, 10</sup> That the Ni steels are so often taken as the standard whose properties it is desired to match is strong evidence of the value of Ni as an alloying element.

**Nickel Steel Castings.** It was pointed out in Chap. 3 that, instead of using plain Ni or plain Mn cast steel, some combination of both, such as 0.30 C, 0.40 Si, 1.00–1.25 Mn, 1.25 Ni with or without 0.10–0.15% Mo, is very often used. Armstrong<sup>11</sup> made a study of heavy castings for Navy use and concluded that the improved properties justified the use of some Ni.

**Plain Nickel.** Even when the Mn is at the usual level, it is unusual to employ more than 2% Ni in cast steel, and when a higher Ni is



S.A.E. 2315, water quenched from 1450-1500°.

S.A.E. 2512, water quenched from 1425-1475°.

Fig. 8.—Properties of low C, Ni steels. Small sections, 1/2- to 1 1/2-in. diam. water quenched and drawn as shown. (Nickel Alloy Steels, 5)

Composition, %

	C	Mn	Ni
2315—	0.10	0.20	0.30
2512—	0.17 max.	0.60	3.75
			4.75-5.25

See general comment on property charts, p. 40, chap. I.

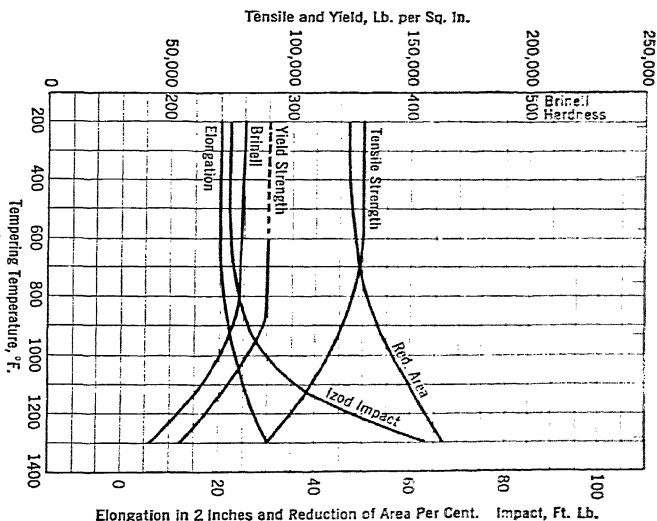
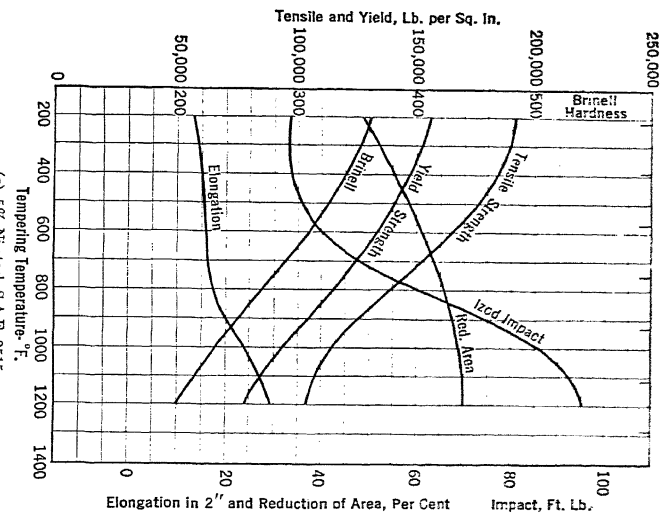


Fig. 9.—Properties of 5% and 1% Ni steels.

See general comment on treatment chart on page 123.

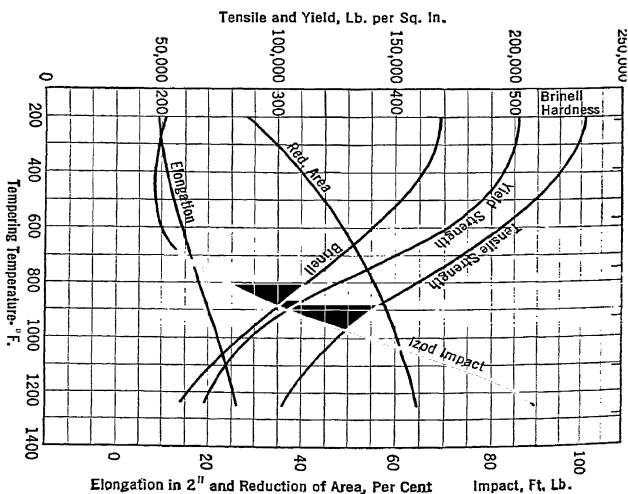
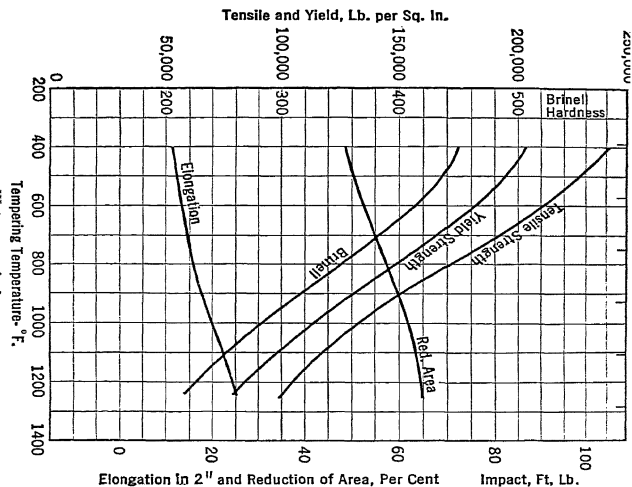
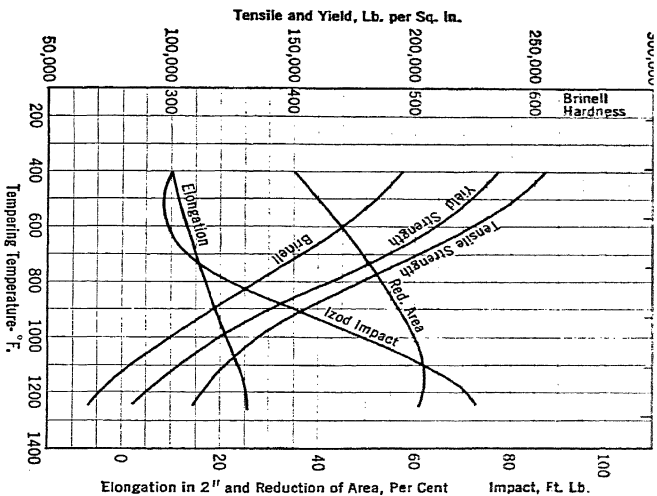


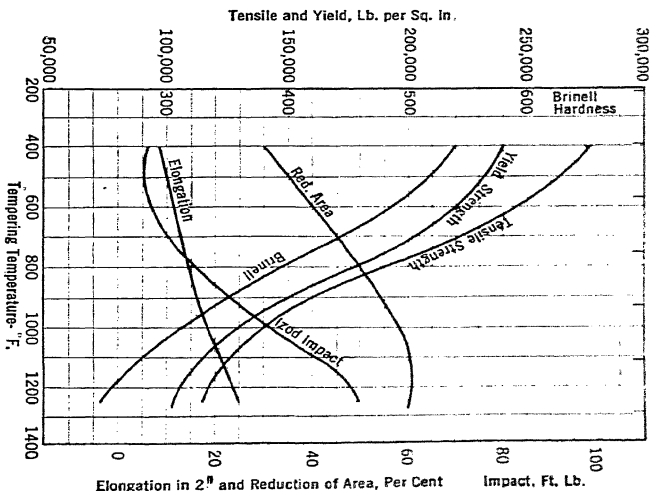
Fig. 10.—Properties of 0.30 C, 3.50% Ni steel, in small sizes,  $\frac{1}{2}$ - to  $1\frac{1}{2}$ -in. diam., quenched from 1325-1475°, drawn as shown.  
(Nickel Alloy Steels,<sup>6</sup>)

See general comment on property charts, p. 40, chap. 1.



S.A.E. 2340.  
Composition, %  
C, 0.35-0.45  
Mn, 0.60-0.90  
Ni, 3.25-3.75

Oil quenched from 1400-1450° drawn as shown.



S.A.E. 2350.  
Composition, %  
C, 0.45-0.55  
Mn, 0.60-0.90  
Ni, 3.25-3.75

Fig. 11.—Properties of the higher C, 3.50% Ni steels, oil quenched, in small sizes,  $\frac{1}{2}$ - to  $1\frac{1}{2}$ -in. diam.

The properties of 0.35 C, 3.50% Ni steel, oil quenched, are intermediate between those for 2330 (Fig. 10) and 2340; those for 0.45 C, 3.50% Ni, between those for 2340 and 2350.

See general comment on tempering steels, p. 11, above.

used, the C is decreased below the usual level since the high-C, high-Ni combination has too much hardening power. Examples of this are given below:

C	Si	Mn	Ni	Norm. from	Drawn at	Tensile	Yield	Elong. %	R.A. %	Impact
0.35	0.31	0.82	3.00	1435°	1100°	111,000	72,000	13	22	
0.20	0.36	0.65	2.40	1725°, 1510°	1200°	76,000	54,500	37.5	59.5	34.5† Charpy
0.17- 0.19	0.40	0.80- 0.95	2.10- 2.25	1775°, 1525°	1100°	79,000- 84,000	50,000- 53,000	28-32.5	59-62.5	55-65‡ Izod

The last two, with their low C, are extremely tough steels.

† Aldridge and Shepherd <sup>12</sup>—Impact at -75°, 22.5 Charpy.

‡ Armstrong. <sup>13</sup>

**Nickel-Vanadium Castings.** A more common composition is 1.40-1.60% Ni with 0.08-0.12% V, which gives:

	Si	Mn	Ni	Norm. from	Drawn at	Tensile	Yield	Elong.	R.A.	Izod	
0.26- 0.32	0.40	0.80- 1.00	1.50- 1.65	0.08- 0.12	1725°, 1525°	1100°	{ 90,000 97,000	{ 60,500 70,500	25.5-28	51-60	47-60*
0.18- 0.25	0.40	0.60- 0.75	1.45- 1.60	0.08- 0.12	1750°, 1525°	1100°	{ 78,500 84,000	{ 57,000 64,000	26-30	47-63	

\* Armstrong. <sup>11</sup>

One might expect that an analogous Ni-Ti steel would have similar properties. According to "Vanadium Steels and Irons," <sup>14</sup> the first steel may be single normalized from 1550-1600° and drawn without materially altering the properties. Double annealing at 1750° and 1475° gives 87,500 tensile, 58,000 yield, 29.5% elong., 55% R.A., 57 Izod. If normalized from 1775°, water quenched from 1500°, and drawn at 550 to 1250°, properties may be obtained, in 1-in. section varying from 150,000 tensile, 110,000 yield, 10% elong., 38% R.A., 17 Izod to 100,000 tensile, 57,000 yield, 27% elong., 58% R.A., 53 Izod, in a steel with both Ni and Mn at the high end of the range.

**Nickel-Molybdenum Castings.** Molybdenum accentuates the air-hardening propensities of Ni steels. The Ni-Mo steels thus require tempering. Zima <sup>15</sup> reports the following:

C	Mn	Ni	Mo	Annealed at	Norm. from	Drawn at	Tensile	Yield	Elong. %	R.A. %
0.30- 0.35	0.60- 0.80	1.25- 1.60	0.30- 0.35	1850°	1700°	1200-1350°	{ 86,000- 90,000	{ 59,000- 62,500	24-26.5	51-61



**Other Nickel Cast Steels.** Data for Ni-Cr and Ni-Cr-Mo cast steels with 1.25 to 3.50% Ni, 0.60–1.00% or even higher Cr, and, if used, 0.20–0.50% Mo are on record. They are sensitive to rate of cooling after normalizing. To develop the full potentialities of these steels liquid quenching, or individually cooling the castings in air to give an air quench, always followed by a draw, is required. These steels will be mentioned in Chap. 12 as they are too highly alloyed to be considered from the point of view of Ni alone.

**Summary.** The outstanding property of Ni is its ability to toughen steel, the increase in strength due to it being gained with less loss in ductility than with most other alloying elements. It enters ferrite only and hence is often accompanied by carbide-forming elements.

It lowers  $A_{c3}$  markedly; so Ni steels are heated to lower temperatures for quenching, etc., than their Ni-free counterparts. It slightly retards grain coarsening. It does not form a gamma loop.

It has strong effect in making austenite sluggish, lowering the  $A_r$  points and the critical cooling rate, and increasing depth hardness. It does not stabilize the carbide, does not make martensite and secondary troostite reluctant to temper, and confers little if any degree of high-temperature strength.

It has a favorable effect upon atmospheric corrosion resistance, in this respect resembling Cu, but is seldom used primarily for this purpose. An outstanding specific effect is that the impact resistance of Ni-containing steels is retained even at extremely low temperatures.

The toughening effect of Ni is utilized in a wide variety of complex steels in which other elements are used either to enter the carbide, or to replace part of the Ni in ferrite, the replacement being sometimes to secure definite properties, but very often merely to supplant part of the Ni by a cheaper element. Only the first type, i.e., steels with Ni and Cr, Ni and Mo, or Ni, Cr, Mo steels are shown in the S.A.E. lists. Substitution, especially by Mn, is, however, very common in cast steels.

According to a producer of alloy steels, the tonnage of bar stock containing Ni is approximately equal to that of all the Ni-free alloy steels put together (Cu-bearing steel for atmospheric exposure not being included in the comparison). Of the Ni-containing steels, only one-fourth are plain Ni, another one-fourth contain both Ni and Mo, and one-half contain Ni and Cr, with or without Mo or V.

The essentials of the behavior of the Ni and Ni alloy steels are given in "Nickel Alloy Steels,"<sup>5</sup> prepared by French and Sands, and The Alloys of Iron Committee has in preparation a monograph on the

"Alloys of Iron and Nickel" in two volumes, which will deal with the subject in great detail.

## BIBLIOGRAPHY

1. MAURER, E., and H. KORSCHAN: Beitrag zur Kenntnis der mechanischen Eigenschaften grösster Schmiedestücke (Study of the Mechanical Properties of Very Large Forgings), *Stahl u. Eisen*, v. 53, 1933, pp. 209-215; 243-251. 271-281.
2. American Welding Society: "Welding Handbook." New York, The Society, 1938, 1210 pages.
3. GRIZZONI, G.: "Gli Acciai." Milan, u. Hoepli, 1932, 471 pages.
4. MÜLLER-HAUFF, A., and K. STEIN: "Automobile Steels." New York, J. Wiley & Sons, Inc., 1930, 219 pages.
5. FRENCH, H. J., and J. W. SANDS (editors): "Nickel Alloy Steels." New York, International Nickel Co., Inc., 1934, 7 sections.
6. SMITH, S. D.: An X-ray Study of the  $Ar_3$  and  $Ac_3$  Points of Iron and Iron-Nickel Alloys, *Am. Soc. Metals*, October, 1937, Preprint 19, 8 pages.
7. ANGELL, W. R.: Mechanical Properties of Some Structural Steels Hardened Below the  $Ac_3$  and Above the  $Ar_3$  Point, *Trans. Am. Soc. Steel Treat.*, v. 17, 1930, pp. 262-272.
8. JONES, J. A.: The Present Trend in Alloy Constructional Steels, *Iron Steel Ind.*, v. 10, 1936, pp. 194-201.
9. HOUDREMONT, E., and H. KALLEN: Schwachlegierte nickel-freie Baustähle (Low Alloy Nickel-free Structural Steels), *Tech. Mitt. Krupp*, v. 2, 1934, pp. 117-126.
10. HAIN, E.: Die Nickelfreien legierten Baustähle für den Kraftfahrbau (Nickel-free Alloyed Structural Steels for Construction in Transportation, *Automobiltech. Zeit.*, v. 40, 1937, pp. 299-304.
11. ARMSTRONG, T. N.: Properties of Some Cast Alloy Steels, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 286-318.
12. ALDRIDGE, B. G., and G. C. SHEPHERD, JR.: Nickel Steels at Low Temperatures, *Metals & Alloys*, v. 7, 1936, pp. 147-152, 185-192.
13. ARMSTRONG, T. N.: Nickel Steel Castings in Railroad Rolling Stock, *Metal Progress*, v. 33, February, 1938, pp. 163-166.
14. Vanadium Corporation of America: "Vanadium Steels and Irons." New York. The Author, 1937, 189 pages.
15. ZIMA, A. G.: Properties and Uses of Some Cast Nickel Alloy Steels, *Trans. Am. Foundrymen's Assoc.*, v. 41, 1933, pp. 199-224.
16. ERSTEIN, S., J. H. NEAD, and J. W. HALLEY: Choosing a Composition for Low-Alloy, High Strength Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 120, 1936, pp. 309-345. Also: *Metals Technology*, v. 3, 1936, Tech. Publ. 697, 31 pages.

## CHAPTER 7

### THE PLAIN CHROMIUM STEELS

**General Behavior.** *Cost and Slag Losses.* According to Kinzel and Crafts<sup>1</sup> more Cr is used in steel than any other of the added metals, with the exception of Mn and Si which are, of course, used for de-oxidation as well as for alloying effect. A primary reason for this large use as alloying element is the cheapness of Cr. When it may be added as high-C ferrochromium, Cr costs but about 10½ cents per lb. of contained Cr, at present prices. If, as in some of the low-C, corrosion-resistant steels, it must be added as the low-C ferro-alloy, the cost approximately doubles. It is more readily oxidized in melting than Fe, so there is a Cr loss in remelting scrap, and the entrance of Cr oxide into the slag somewhat complicates the steel-making process.

110

P. 90

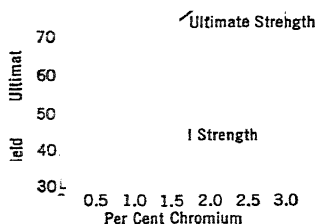


FIG. 1.—Effect of Cr in normalized low C steels. (Epstein, Nead and Halley.<sup>18</sup>)

**Chromium Enters Both Ferrite and Carbide.** Chromium functions as a ferrite strengthener, as is shown in Fig. 1, but in even more marked fashion, as a carbide former. It confers depth-hardening properties and tends to restrain grain growth. In the absence of C, several per cent of Cr will dissolve in ferrite. In the presence of C, Cr divides between ferrite and carbide. The cumulative strengthening effect of Cr and C is shown in Fig. 2.

Dissolved either in ferrite or austenite, it gives the steel ability, when in an oxidizing environment, to form a protecting oxide film on the surface, this ability increasing with Cr content, and becoming very marked at about 12% Cr. This is a very specific effect of Cr. The austenitic steels will be discussed in Chap. 20.

Chromium-bearing cementite and complex carbides of Cr plus certain other metals are very hard. Chromium steels with free carbides are extremely wear resistant and suitable for cutting edges. This specific effect is an outstanding characteristic of Cr steels.

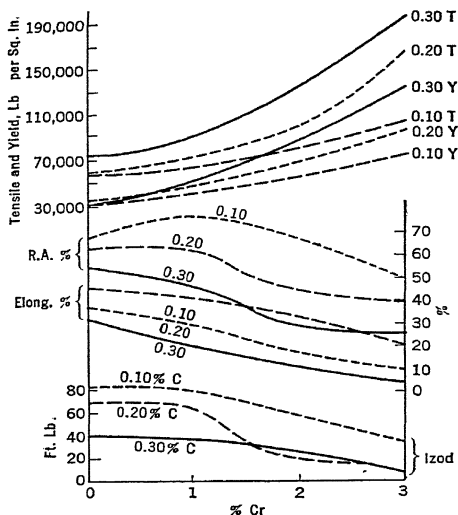


FIG. 2.—Mutual effect of Cr and C in normalized steels of 0.10, 0.20, and 0.30% C, with 0.30–0.60% Mn and varying Cr. Small sections. After Union Carbide and Carbon Res. Labs.

See general comment on property charts, p. 40, chap. 1.

**Range of Composition.** This combination of properties makes Cr useful in a very wide range of composition, not only in combination with the other alloying elements, but even in plain C-Cr steels. The C-Cr steels comprise not only constructional, but a wide variety of special-purpose steels. Commercial uses are met for steels with as little as 0.25% or less Cr, and for heat-resistant alloys with as much as 35% or more, and with a considerable range of C contents. The Cr:C ratio varies according as the aim is to utilize the ferrite-strengthening or the carbide-hardening tendency of Cr.

**The Gamma Loop.** Chromium forms a gamma loop in the equilibrium diagram. The position of the loop is profoundly affected by

C content. Fig. 3. Hardening by quenching is impossible at compositions to the right of, and wholly outside, the gamma loop.

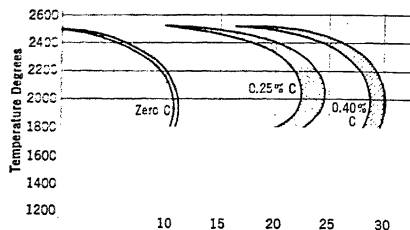


Fig. 3.—Gamma loops, for steels varying in C and Cr.

Schematic, after Bain.<sup>4</sup> Gamma lies within the loop, gamma plus alpha within the band, and alpha outside the loop. A steel to the right of the loop representing its C and Cr content, does not harden on quenching.

**Matters Affecting Heat Treatment.** The eutectoid C content is lowered by Cr (Fig. 4), the  $A_{c3}$  temperature is raised by Cr (Fig. 5). The  $A_{cm}$  line is straightened up and much higher temperatures required

127—

0 0.2 0.4 0.6 0.8 1.0  
% C

Fig. 4.—Eutectoid composition in Cr steels. (After Monypenny.<sup>15</sup>)

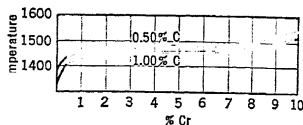


Fig. 5.—End  $A_{c3}$  in Cr steels. (After Maurer and Nienhaus.<sup>19</sup>)

to put all the carbide into solution in the austenite (Fig. 6). Even at temperatures in the austenite field where all carbide is truly soluble, the rate of solution is generally held to be low. Undissolved carbides are postulated as tending to inhibit grain growth, and to the low rate of solution is ascribed the tendency of Cr steels to remain fine grained on heating, unless both sufficient temperature and sufficient time are allowed for complete solution of carbide and homogenization of austenite. On the whole, addition of Cr makes a steel require higher heating temperature for heat treatment.

Since homogenization of austenite is required to produce maximum sluggishness as to transformations on cooling, the depth-hardening

properties of Cr steels are sensitive to the heating temperature and time. The air-hardening properties are therefore affected by C and Cr composition and also by the maximum temperature and the time of heating. The structures of normalized steels are indicated in Fig. 7.

**Carbides.** The cementite in the pearlite referred to in Fig. 7 is not pure  $\text{Fe}_3\text{C}$ ; it carries some Cr replacing Fe, i.e., it is a Cr-bearing cementite. With sufficiently high C and Cr, a new carbide appears. This has a different crystal structure from cementite, and a different formula. It is trigonal in crystal habit and has the formula  $\text{Cr}_7\text{C}_3$ .

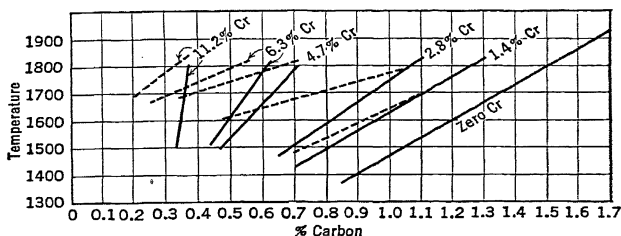


FIG. 6.—Carbide absorption in the heating of Cr steels.

Solid lines ( $A_{cm}$  lines), show disappearance of hypereutectoid carbide. Dashed lines show disappearance of pearlitic carbide. (After Monypenny.<sup>15</sup>)

This can carry some Fe replacing Cr. At still higher Cr content (without the C necessarily being high) a third carbide appears, this having the formula  $\text{Cr}_4\text{C}$ . Like the other carbides, this can carry Fe replacing Cr. Thus, depending on C and Cr content, temperature of heating and rate of cooling, a Cr steel may contain  $(\text{Fe,Cr})_3\text{C}$ , cementite, which is mostly Fe carbide but may contain up to 15% Cr;  $(\text{Cr,Fe})_7\text{C}$ , trigonal carbide, which may contain up to about 50% Cr; or  $(\text{Cr,Fe})_4\text{C}$ , cubic carbide, which contains mostly Cr, very little Fe. Investigators are not in exact agreement as to the compositions of these carbides. The mixed Fe and Cr carbides will vary in their solubility in austenite according to which carbide predominates in the mixed carbide, and on annealing the mixed carbide may decompose, part staying in solution and a carbide of a different character separating out. These changes are of course influenced by the presence of other alloying elements, when other carbide formers are present (such as W or Mo which can themselves form carbides, and can replace Fe or Cr in cementite or in Cr carbides, to form still more complex mixed carbides), so the formation, solution, and separation of any particular carbide whose chemical composition can be clearly defined becomes unpredictable.

It will therefore be necessary to refer in rather general terms, in

discussing Cr and complex Cr steels, to the effect of C or to the behavior of carbide rather than to connect the various phenomena with a particular carbide. It should be recalled, however, that in the higher C and the higher Cr steels, the carbide may be of quite different nature from the cementite of the constructional steels.

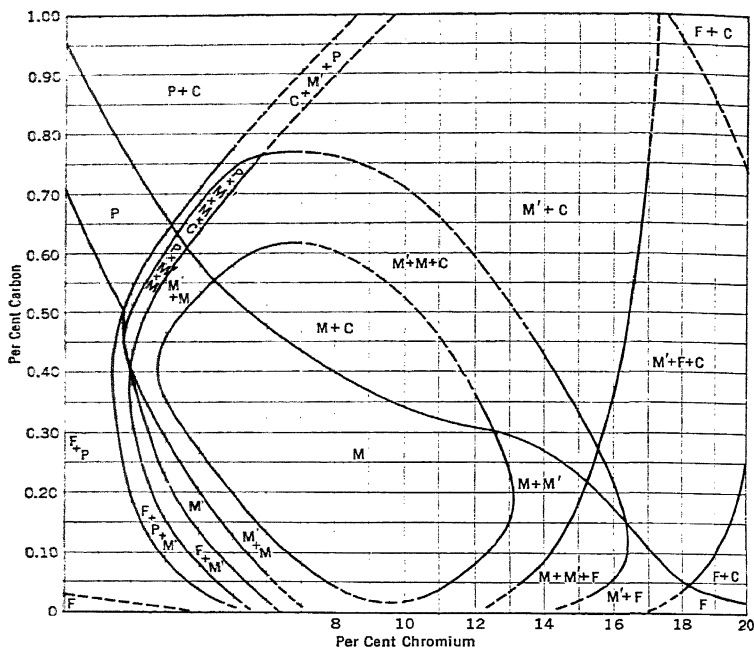


FIG. 7.—Constitutional diagram of Cr steels containing 0.50% Mn, normalized in 1 in. diam. bars. F, ferrite; C, carbide; P, pearlite, M', pseudomartensite; M, martensite. (Courtesy Union Carbide and Carbon Res. Labs.)

Those steels analogous to ordinary ferritic, pearlitic, and cementitic steels and designed for treatment by quenching and tempering contain low Cr, usually under 1.75%. The greater part of this chapter will be devoted to such steels. Certain tool, die, and special steels, to be discussed in later chapters, utilize the hardness of the special carbides.

**Intermediate Chromium Steels.** There is an important class of steels used for cracking tubes and the like in the oil industry, which

contain low C and from about 2 to about 9% Cr, for oxidation and corrosion resistance. These are especially sensitive to rate of cooling. Figure 8 shows data from Wright and Mumma<sup>2</sup> for properties of normalized and of annealed steels of this group.

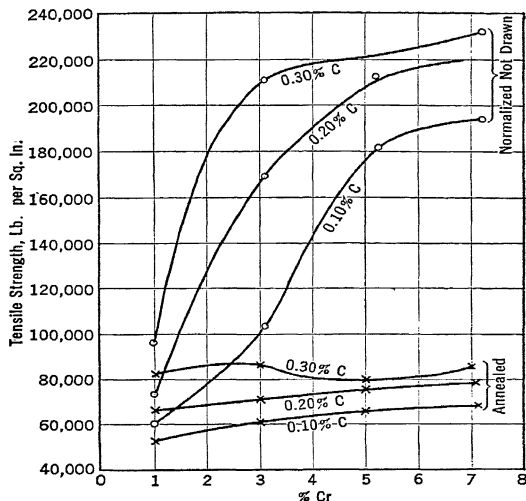


FIG. 8.—Sensitivity of intermediate Cr steels to cooling rate.

The 0.10 C, 1.00% Cr and 0.10 C, 3.00% Cr steels were normalized from or annealed at 1650°. All others from or at 1600°. Specimens 0.675 in. diam. (After Wright and Mumma.<sup>2</sup>)

A large increase in hardenability occurs at about 2.50% Cr. Kinzel and Crafts<sup>1</sup> suggest that this may be coincident with the appearance of trigonal carbide.

**Critical Cooling Rate.** The effect of Cr on decreasing the critical cooling velocity has been studied by Esser, Eilender, and Majert.<sup>3</sup> Working with 0.35–0.55% C steels, they find the critical cooling rates, to avoid formation of pearlite, in steels quenched from 1740° to be:

CR CONTENT, %	DEG. PER SEC.
none	1000
0.5	725
1.1	180
2.0	40
2.6	15–30

Most air-hardening steels utilize Cr, in conjunction with other elements, because of this effect. As Kinzel and Crafts<sup>1</sup> point out, the



amount of Mn in a Cr steel markedly influences the critical cooling rate.

The hardened Cr steels require higher tempering temperatures for an equivalent degree of softening than do C steels.<sup>4</sup> Related to this is a mild strengthening effect of Cr in tests at high temperatures, though other elements are usually added to accentuate this effect.

**The Toughness Problem.** Chromium has not had much acceptance in amounts over 1% in plain C structural steels for heat treatment since it is considered that strengthening due to further increase is at the expense of toughness. However, Wright and Mumma,<sup>2</sup> some of whose data are shown in Fig. 8, have shown that this is not true if the C is held down; and Aitchison<sup>5</sup> has shown that even medium-C, rather high-Cr steels display good toughness if drawn at a sufficiently high temperature. Jones<sup>6</sup> concluded that if the Mn or other hardening elements were held low, a 0.30–0.35% C steel could utilize 2.25–3% Cr, and one of 0.35–0.40 C, 1.5 to 2.25% Cr. Other authorities, whose work is conveniently summarized by Greaves,<sup>7</sup> indicate that the maximum properties are secured at about these proportions in quenched and tempered steels. Since depth hardening increases with the Cr content, such steels might be desirable for large sections. Indeed, in spite of the fact that Fig. 7 indicates that they would normalize to an excessive hardness, the low- and medium-C normalized 3% Cr steels show good ductility for their high strength, and, by drawing, high toughness may be produced.

**Chromium Steels Usually Complex.** Ordinarily, in constructional steels, the Cr is kept on the low side and other alloying elements used with it. The combination of Cr with another carbide-forming element is usually more effective than increasing the Cr alone, while the combination of Cr with a ferrite-strengthening, toughening alloy often makes for the desired balance of hardness and toughness. The complex Cr steels are, therefore, more commonly used for structural purposes, and in many varieties of tool steels Cr enters as one element among several.

**S.A.E. Chromium Steels.** The S.A.E. list includes the following plain Cr constructional steels:

	% Cr		% S		
5120	0.15–0.25	0.30–0.60	0.60–0.90	0.05 max.	0.04 max.
5140	0.35–0.45	0.60–0.90	0.80–1.10		
5150	0.45–0.55	“	“		
52100	0.95–1.10	0.20–0.50	1.20–1.50	0.035 max.	0.03 max. (A tool steel)

and the corrosion-resistant series:

	% C	% Mn	% Cr
51210	0.12 max.	0.60 max.	11.5-13
X51410	"	"	13 -15
51510	"	"	14 -16
51710	"	"	16 -18
51335	0.25-0.40	"	12 -14 (cutlery steel)

**Other Commercial Steels.** The 4 to 6% Cr steels with low C, as well as other steels of higher and lower Cr, intermediate between the two S.A.E. classes used for tubes, etc., in the petroleum cracking industry are not in the S.A.E. list. Modern practice generally adds Mo to these intermediate steels for such purposes. The properties, resistance to oxidation and to corrosion by oil, plus load-carrying ability at elevated temperature, are due to both elements. These Cr-Mo steels will be discussed in Chap. 19.

The production of S.A.E. bar stock of plain Cr and complex constructional steels containing Cr is given by one producer as being about equal to the production of plain and complex Ni-steel, i.e., 6% each of plain Cr and plain Ni, 6% each of Cr-V and Ni-Mo, and 12% of Ni-Cr and more complex Ni-Cr compositions. The percentage of Cr in these steels, however, averages lower than Ni.

**Chromium Carburizing Steels.** Since low-C steels with Cr have toughness and high-C steels with Cr have hardness and wear resistance, the use of low-C steels with Cr for carburizing is logical. Moreover, Cr aids in speed of carburizing and tends toward the formation of a higher C case than is met with in other alloying elements. Discussion of these steels will be deferred until Chap. 14.

**Low-Chromium Constructional Steels.** *Effect of 0.25 Chromium.* Although it is usually considered that upward of 0.50% Cr is needed to secure much benefit from its addition, McQuaid \* has pointed out that, especially in the medium Mn steels of the fine-grained type (where the depth-hardening effect conferred by the Mn is reduced by making the steel fine grained), the hardenability can be restored by substitution of around 0.25% Cr for that much Mn. He further states that warpage is decreased, and credits this small amount of Cr with beneficial effect in medium-C and oil-hardening steels, including 0.80 C, 0.80% Mn spring steel.

\* See "The Alloys of Iron and Chromium," vol. I, p. 157.

This is a good illustration of the fine gradations possible in alloy steel development. From plain C steel one goes to increased Mn as a cheap alloying element, makes this fine grained to produce the benefits mentioned in Chap. 9, Vol. I, but in so doing loses some of the hardening effect of Mn, then builds up the hardening power with a little Cr. Such adjustment never creates an exact duplicate of another composition, but rather a steel of definite individuality—in this case the improved resistance to warping reflects the individuality of this particular steel.

**Effect of 0.50% Chromium.** In Cr-C steels of medium C content, the addition of 0.50% Cr will give, after heat treatment and dependent upon the size of section, an increase of 15 to 20% in tensile and yield strengths, with small loss in ductility, over straight C steels of the same C and Mn content. This is illustrated by the following. A and B are on 1-in. bars quenched in oil and drawn at 1200°. C is from machine-gun barrels quenched in oil from 1600° and drawn for two hours at 1225°.

	Mn	Cr %	Tensile	Yield	Elong. %	R.A.
A 0.35	0.51	nil	84,000	61,500	29	71
B 0.36	0.44	0.57	103,200	76,000	27	62
C 0.51	0.81	0.49	111,000-120,000	80,000-87,000	22.5-23.5	59.5-63.5

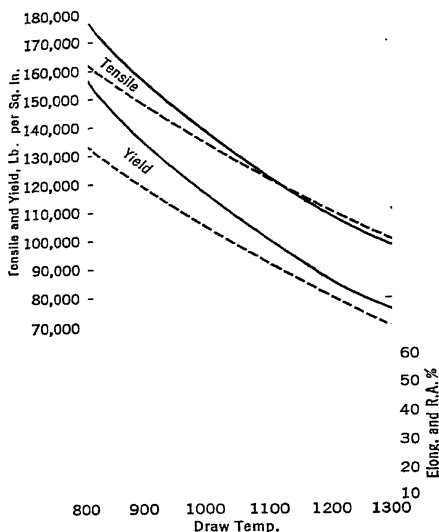
In the hardened condition these steels give excellent service for wearing surfaces, such as small gears, jaws for wrenches, etc.

**As-rolled Steels.** For larger sections which are not heat treated or annealed after rolling or forging, the addition of 0.50% Cr increases the tensile and yield strengths by some 5 to 10%, and gives the ductility of a C steel of some 10 points lower C content. This is illustrated by the following examples:

	C %	Mn %	Cr %	Tensile	Yield	Elong. %	R.A. %
4-in. . . . .							
1 billet . . .	0.40	0.50	0.55	93,000	72,000	26	50
8-in. . . . .							
1 bar . . . .	0.47	0.55	0.57	95,000	50,000	22	50

**Effect of 1% Chromium, S.A.E. 5100 Series.** The effect of Cr may be compared with that of Mn, which we have already appraised as a potent strengthening element, in Fig. 9. With equivalent ductility the

tensile and yield strengths are higher, in oil-quenched and drawn steels, upon the addition of 1% Cr than upon the addition of 1.75% Mn.



Solid lines, 5140.  
 Normalized from 1675°.  
 Oil quenched from 1550°.  
 Drawn as shown.  
 Dashed lines, T 1340.  
 Normalized from 1650°.  
 Oil quenched from 1500°.  
 Drawn as shown.  
 Small sections.  
 The R.A. curves coincide.

FIG. 9.—Comparison of 1% Cr (5140) and 1.75% Mn (T 1340) steels. (S.A.E. curves.)

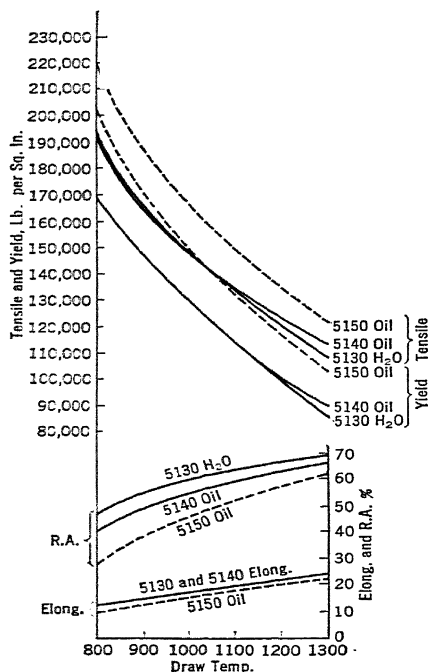
See general comment on property charts, p. 40, chap. 1.

At constant Cr content, addition of C strengthens materially and reduces ductility. Very high strengths are obtainable in small sections by oil quenching in the higher C contents, as Fig. 10 shows.

**Complex Chromium Steels.** Probably on account of the strong carbide-forming tendency of Cr it is not used indiscriminately to replace other alloying elements in complex alloy steels of medium and high C. It is necessary to “balance” such constructional steels containing Cr lest they lose toughness. In the Ni-Cr series, for example, the Ni is used in larger percentage than the Cr. Other ferrite formers may be used to balance the Cr, e.g., Si and Cu, both of which combinations have utility. Care has to be taken in blending Cr with Mn because the latter is not solely a ferrite former, and the combination of the two carbide formers is likely to lend hardness without toughness unless the C is kept low. Chapters 12 and 13 will deal with complex steels.

**Higher Chromium and Higher Carbon Steels.** In the lower ranges of C and Cr the effect of Cr is rather analogous to that of Mn. As we depart from the composition range of the constructional type of steels and raise either C or Cr, special properties due to Cr begin to appear.

**Tool Steels.** With increase in C the influence of Cr becomes marked, owing to the increasing amounts of double carbide. The hardness increase is greater proportionally than the C increase. For well-bits and jars in the hardened condition, Cr tool steel has no equal among the low-priced alloys or straight C steels. For die-blocks used in drop-forging work it does not seem quite to "hit the mark," ap-



5130, 0.30% C, normalized from 1650°,  
water quenched from 1550°  
5140, 0.40% C, normalized from 1625°,  
oil quenched from 1550°  
5150, 0.50% C, normalized from 1625°,  
oil quenched from 1550°

Drawn as shown.

(Bethlehem Alloy Steels.<sup>20</sup>)

FIG. 10.—Effect of C in 0.80-1.10% Cr steels.

See general comment on property charts, p. 40, chap. 1.

parently not having the requisite toughness to offset the brittleness, especially in the larger sections. With 0.70 to 0.80% C it makes an excellent chisel, and with 0.90 to 1.00% C and about 0.60% Mn it gives even better results for pneumatic chipping chisels than do many varieties of high-alloy steel.

Use has been made of 0.50% Cr steels for tools, such as drills, saw-blades, knives, razors, files, and other tools requiring a keen cutting edge. The function of Cr is to give greater surface hardness and

greater depth of hardening, rather than direct increase in cutting ability.

**One to 2 Per Cent Chromium: Bearing Steels.** Chromium steels with about 1.00 to 1.75% Cr, with high C, find their greatest use in balls, ball-races, cones, roller bearings, crushing machinery, safe steel, tools, and other parts requiring a very hard surface. The use of about 1% C combined with the above Cr content appears in general to give the highest combination of "tough-hardness" and plasticity; it is customary, however, to vary the C and Cr contents according to the mass-surface factors of the definite piece involved, and, in turn, to the depth of hardening required. Steel of this type tends to have a coarse ingot structure, which should be broken down on forging. Moderately slow

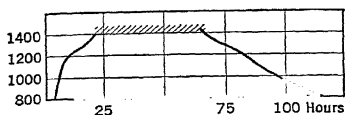


Fig. 11.—Annealing chart, 1.10 C, 1.75% Cr steel.\*

cooling in annealing tends to separate excess carbide at grain boundaries, but not to break up the pearlite. To make it readily machinable, the eutectoid carbide as well as the excess carbide must be spheroidized. This may be accomplished by a very long heating at a temperature in or very slightly above the critical range, followed by a very slow cooling. The elements of time, mass, and surface enter very strongly into such annealing. The temperatures to be used for steel for ball and roller bearings having a chemical composition of 0.90 to 1.10% C and 1.25 to 1.75% Cr will vary between 1400 and 1425°; the time of saturation will depend upon the mass and exposed surface of the charge and usually will be several days. Such steel in the soft annealed condition should have a Brinell hardness of not over 170. The annealing cycle of this steel, to give a Brinell hardness of 149, is shown in Fig. 11.

An expeditious method of annealing this steel is to normalize and then anneal, as follows: first, thoroughly heat the steel at a temperature above  $A_{c_m}$  for the solution of the cementite; air cool beneath Ar to prevent accumulation of carbides at grain boundaries; reheat to a temperature in or slightly over the  $A_c$  range to refine the grain; and slowly cool in the furnace or in lime to obtain maximum softness. For a steel containing 0.90 to 1.10% C, under 0.50% Mn, and about 1.00% Cr, the following temperatures may be used to advantage:

1. Heat to 1700 to 1750°.
2. Air cool to about 800°.
3. Heat to 1400 to 1425°.
4. Slow cool in furnace or in lime.

\* Wallquist and Larsson (Jernkontorets Ann. v. 22, Dec. 1938, pp. 611-46) report that steel of 1% C, 1.5% Cr may be annealed in 9 hours by using a specified sequence of temperatures.

**Hardening.** When hardened in water or oil, these steels take on great hardness, both on the surface (about 100 scleroscope) and at depth. The critical ranges for these steels, with over 0.90% C, will vary from 1360 to 1375° for 0.50% Cr; and from 1400 to 1425° for 1.50% Cr. The Cr magnet steels will be discussed in Chap. 24.

**Steels of Higher Chromium Content.** The sensitivity of steels of 3 to 7% Cr to variations in cooling rate has been brought out in Fig. 8. This behavior needs further attention as we go to still higher Cr.

Some of the plain Cr steels are heat treatable, some are not, according to the C content. The ability for heat-treatment of course means that the alpha-gamma change occurs so that pearlite may form from austenite or that by quenching through Ar' we may secure the formation of martensite at Ar'.

In carbonless Fe-Cr alloys the austenite field does not extend beyond about 14% Cr (Fig. 3). An alloy with 15% Cr is alpha iron at all temperatures in the solid state. It cannot be hardened by quenching or grain refined by any heat treatment, whereas one of say 9% Cr can be.

As C is added, the austenite field (gamma loop) is extended as Fig. 3 shows.

If we look at the Fe-Cr-C system from the point of view of the effect on C steel of the entrance of Cr, a peculiar situation is found. This is shown in Fig. 12.

In the figures the fields in which alpha plus gamma iron exists are cross hatched. At the top of the Fe-C diagram is a small field for what in C steel is called "delta" iron, but which is merely alpha formed as the metal freezes, then changing to gamma at lower temperatures, again to become alpha at Ar<sub>3-1</sub>.

The introduction of Cr enlarges this alpha, or delta, plus gamma field and moves the eutectoid point to the left. Note at 14% Cr how the upper ("delta") field has come down to meet the lower alpha field. The eutectoid C content has now moved to the left to about 0.35%, so that a 0.40 C, 14% Cr steel is hypereutectoid. At 18% Cr the eutectoid C percentage begins to move toward the right. The curvature of the boundaries for the gamma field show that in an 18% Cr steel at 0.40% C cooling produces gamma + alpha, alpha, gamma + alpha in order and at 0.20% C, alpha, gamma + alpha, alpha. The area to the left of the cross-hatched band is for alpha (delta), that to the right, for gamma iron. Note also that Ac<sub>1</sub> is rising as Cr increases. Figure 13 summarizes the data of Fig. 12.

**Corrosion Resistance.** Many of the uses for steels of 2 to 35% Cr depend on the resistance of such steels to various types of corrosion.

The atmospheric corrosion resistance of Cr steels increases with the Cr content, but 12% or more is necessary to prevent rusting in air completely. Corrosion in nitric acid drops rapidly as Cr increases. The resistance to corrosion in still or evaporator tubes by corrosive

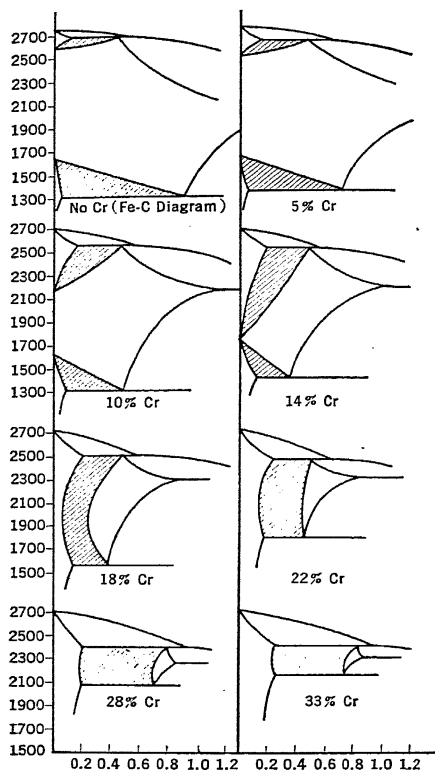


FIG. 12.—Sections of Fe-Cr-C constitutional diagram at successively higher percentage of Cr. (Krivobok and Grossmann.<sup>21</sup>)

The cross-hatched fields are for alpha (or delta) iron, plus gamma iron.

crude oil increases with Cr. Resistance to oxidation and scaling tested at 1275° rises similarly. At still higher temperatures resistance increases with Cr, but the curve is not a smooth one, the 18% alloy will show some scaling, and the use of 30% or more Cr is beneficial. These effects are shown in Fig. 14. Obviously, there are corrosion-resistant uses for Cr steels of as little as 2%, and wider uses for those of 12% and more.



Broadly speaking, the corrosion resistance against oxidizing solutions, in the atmosphere, or against oxidation at high temperatures, is a function of the Cr content and is influenced only in minor degree by the addition of other elements. For resistance to scaling, introduction of Si or Al along with Cr may be helpful, and corrosion resistance to chemical solution is slightly modified by various additions. On the whole, however, the corrosion protection due to Cr can be maintained when additions of other elements are made to control other properties. The steel can be made austenitic by addition of Ni or Mn, the pearlitic

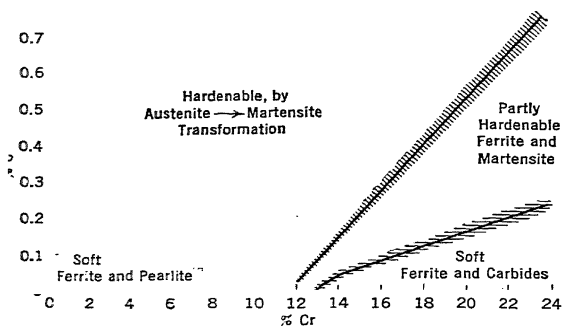


FIG. 13.—Hardenability of FeCrC alloys. (Bain.<sup>22</sup>)

steels can be given high temperature strength by addition of Mo or W, and many other modifications can be introduced without greatly affecting the corrosion resistance conferred by Cr. It should be stated that the corrosion resistance is against oxidizing conditions only. Strong chloride solutions attack the Cr steels.

**Tube Steels.** The Cr ferrites are particularly useful in tubes for oil-cracking stills which operate at around 800–1100°, and at very high oil pressures. Here oxidation on the outside of the tube must be combated, and at the same time the inside must resist attack by corrosive crudes.

A balance of the value of these two types of corrosion resistance against cost early led to the use of 5% (usually termed "4 to 6") Cr steel of low-C content for still tubes. Later experience led to the use of lower and higher Cr contents for less or more severe operating conditions, so that tubes are now commercially obtainable in the range from 2 to 9%, or higher, Cr. These pearlitic Cr steels, while superior to C steels, still lack load-carrying ability at high temperatures. This is supplied by Mo (sometimes by the more expensive W) so that this

class of still tube steels owes its desirable properties equally to the two alloying elements. The tubes may, in some still designs, have to

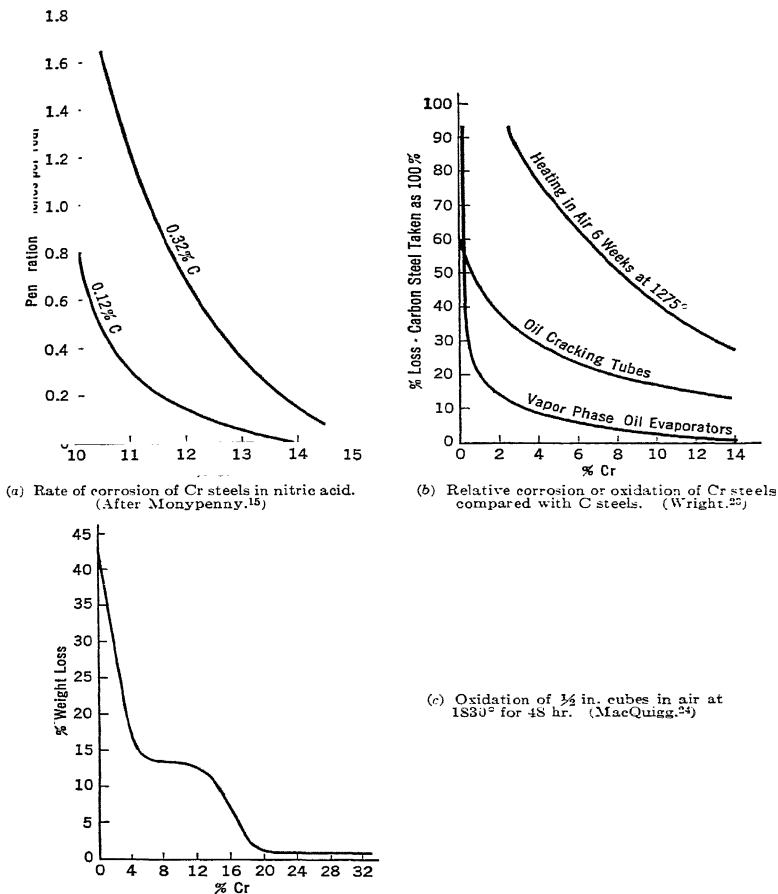


FIG. 14.—Effect of Cr content on corrosion and oxidation resistance.

have their ends rolled into place, like a boiler tube, so are supplied soft annealed and at least partly spheroidized. The room temperature strength for annealed 0.15 (max.) C, 2, 5 or 9% Cr, with or with-

out about 0.50% Mo in the two lower and 1.50% Mo in the 9% Cr. ranges from 65,000 to 80,000 tensile, 25,000 to 35,000 yield, 30 to 35% elong., and 50 to 70% R.A., according to C and alloy content.

Even with the low-C contents common in these steels, and even at only 2% Cr. there is some air-hardening tendency, which is of course very marked in the 9 Cr, 1.5% Mo. Without Mo the steels are prone to grain coarsening and hence to embrittlement in still tube service. The embrittlement, however, does not exist at operating tem-

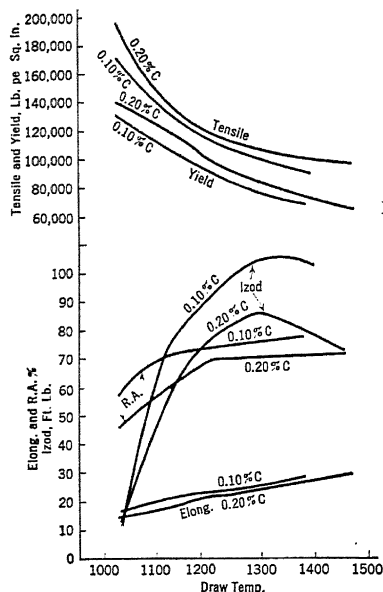


FIG. 15.—Properties of 5.20% Cr steel

(Wright and Mumma.<sup>2</sup>)

0.10% C, normalized from 1600°, drawn as shown.

0.20% C, normalized from 1375°, drawn as shown.

Treated in 0.575 in. diam. bars.

See general comment on property charts, p. 40, chap. 1.

peratures. Under overload at such temperatures they stretch gradually and do not give away suddenly. When the grain has been coarsened and the steel is cooled to room temperature, brittleness ensues, and in cleaning the tubes from coked C, which must be done periodically, they may be broken. The embrittlement and grain growth are prevented by addition of Mo.

Wright and Mumma<sup>2</sup> examined 1 to 7% Cr (see Fig. 8) normalized in 0.675-in. diam. and tempered at temperatures from 1020 to 1380°. Selected data from their list are given in the following table:

C %	Cr %	Draw Temp.	Tensile	Yield	Elong. %	R.A. %	Izod	Charpy
0.10	1	No draw	60,000	44,500	41.5	75.5	88	63
		1225°	57,000	40,500	42	77.5	86	68
		(Annealed)	54,000	32,500	43	77.5	88	60
0.10	3	No draw	102,500	60,500	19	49	30	25
		1200°	80,000	56,000	26	76	98	68
		(Annealed)	60,000	23,500	40	78	88	58
0.10	5	No draw	181,500	109,000	15.5	53	23	18
		1200°	100,000	86,500	23	27	96	51
		(Annealed)	66,000	27,000	37.5	76	77	47
0.10	7	No draw	192,000	111,500	13	42.5	20	21
		1125°	117,000	102,500	20	66	68	41
		1290°	95,500	80,000	24	74	99	53
		(Annealed)	68,000	29,000	37	78	96	67
0.20	1	No draw	73,000	50,500	35	69	77	48
		1250°	70,000	48,000	36	67	82	50
		(Annealed)	66,000	38,500	35.5	60.5	65	41
0.20	3	No draw	168,500	102,500	15.5	40.5	19	18.5
		1225°	104,000	87,000	24	71	87	47
		(Annealed)	72,000	28,000	34	73.5	67	43
0.20	5	No draw	212,500	114,000	9	18.5	23	16
		1125°	137,500	120,000	18.5	59	36	39
		1290°	104,500	83,500	23	68	86	54
		(Annealed)	75,500	32,000	32	75	84	67
0.20	7	No draw	222,500	119,000	10	23	19	16
		1125°	129,500	109,500	20	65	74	38
		1290°	105,500	83,000	24	71	88	49
		(Annealed)	79,000	35,500	33	76.5	104	61
0.30	1	No draw	97,000	62,500	24.5	60.5	42	32
		1200°	93,000	57,000	27	60	41	33
		(Annealed)	81,500	44,000	28	49	39	19
0.30	3	No draw	207,500	116,500	10.5	26.5	19	18
		1215°	123,000	100,500	22	67.5	69	39
		1380°	96,500	74,000	27	72	94	53
		(Annealed)	87,500	37,000	30	65.5	13	28
0.30	5	No draw	221,500	117,000	13.5	31.5	17	15
		1215°	120,000	97,000	21	66.5	76	41
		1380°	99,500	73,500	28	72	77	60
		(Annealed)	79,000	33,500	33.5	75.5	80	79
0.30	7	No draw	233,000	122,500	12.5	35.5	16	10
		1125°	135,000	112,000	19	62.5	62	32
		1200°	123,000	99,000	21	65	73	39
		1380°	104,000	76,500	27.5	70.5	92	65
		(Annealed)	84,500	37,500	33.5	75	90	80

As the table and Fig. 5 bring out, even the 0.10 C, 5% Cr steel is distinctly air hardening. The eutectoid C content with this amount of Cr is only about 0.50%, so what appears to be a low C is medium C when considered in relation to the eutectoid. The tube steels are often specified to contain 4 to 6% Cr with C as desired, ranging from under 0.10 to 0.25% max. C. Figure 15 shows the properties of normalized and drawn steels in this general range.

As a structural material the toughness of a steel of this class is high in relation to tensile strength, which is probably due to the fact that annealing readily spheroidizes the special type of carbides present instead of producing a regular pearlitic structure and that on reheating an especially fine-grained structure is produced.

Alterations in C and Cr contents and additions of Mo will alter the strength and toughness of the hardened and tempered steels in the usual fashion. Molybdenum of course causes resistance to tempering.

The ability to air harden or quench harden can be removed by addition of sufficient Ti, as the following comparison for normalized, not drawn, specimens shows:

			Normal- ized	Tensile	Yield	Elong. %	R.A. %	Izod
0.11 C	5.41 Cr	0.75% Ti	From 1650°	62,000	29,000	44	79	112
0.10	5.20	nil	1600°	181,500	109,000	15.5	53	23
0.10	5.20	nil	(Annealed)	66,000	27,000	37.5	76	27

Fleischmann and Williams<sup>6</sup> report a similar avoidance of hardening by the addition of 0.70% Al.

**Stainless Steels.** For the production of steels that will remain bright in the air a minimum of about 12% Cr is required. With such a Cr content the eutectoid is at about 0.30% C. At 0.08–0.12% C and 11.50–13% Cr heat-treatable, corrosion-resistant steels result. With such C contents, for maximum response to heat treatment, the Cr content has to be held down to the 13% limit, since the tip of the gamma loop is approached at 15 to 16% (compare Fig. 13), and with more Cr than that the steel becomes ferritic and no longer amenable to heat treatment.

In castings from this type of plain Cr steel, Cr is similarly held down to 13–14% max., for grain size tends to be excessive above that content. Normalizing temperatures must be high to produce satisfactory grain refinement. About 1% Ni improves the strength without sacrificing much ductility.

**Turbine Type.** Oil hardening or, in very thin sections, air hardening, say of a 0.10 C, 12% Cr steel, produces a martensitic structure, but the quenched hardness is only around 450 Brinell, and with C decreased to 0.07% it may drop below 300. The material as quenched, however, has around 15–20% elong., 35–55% R.A. The quenching temperature is usually 1750–1850°. Tempering data from Mochel<sup>9</sup> for one of the strongest steels of the group are shown in Fig. 16. Where combined corrosion and wear must be resisted, hardening and drawing up to 600° are employed. It is noteworthy that on tempering much over 900° the properties change very rapidly, the rate of change being

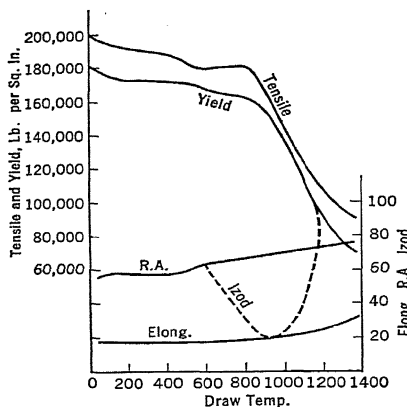


FIG. 16.—Rustless steel, turbine type, 0.09 C, 12.23% Cr, oil quenched from 1750°. Drawn as shown. (Mochel.<sup>9</sup>)

See general comment on property charts, p. 40, chap. 1.

rapid till 1200° is reached. The steels are therefore generally tempered at 1200° in order to secure reproducibility of results. With this draw the steels of this class run 100,000–120,000 tensile, 80,000–100,000 yield, 20–25% elong., 60–65% R.A., 50–90 Izod.

These steels are useful for turbine buckets, valve trim, and other purposes where resistance to superheated steam is required.

Forging of this steel is carried out in the range 1600–1900°. Since the steel is air hardening, the rate of cooling after forging should be slow. Forgings ordinarily require annealing before machining. Process annealing for softening may be at 1450°, full annealing requires soaking at 1550° and slow cooling.

An increase in hardenability is brought about by the addition of a little Ni, so this grade often carries about 0.50% Ni. About 0.50% Mo or V is found to improve workability, and if it is desired to avoid the air-hardening propensity, the C can be bound up by Ti or columbium

(Cb). With the usual amount of other elements for this purpose, i.e., about 6 times the C content for Ti and 10 times for Cb, the steels do not harden on quenching. Instead, the as-rolled steels are softened and toughened by this treatment. This application of Ti is analogous to that already mentioned for 5% Cr steel.

The turbine and analogous services of these steels utilize the strength as well as the corrosion resistance. As sheet and strip they serve for many decorative purposes—automobile hubcaps, for exam-

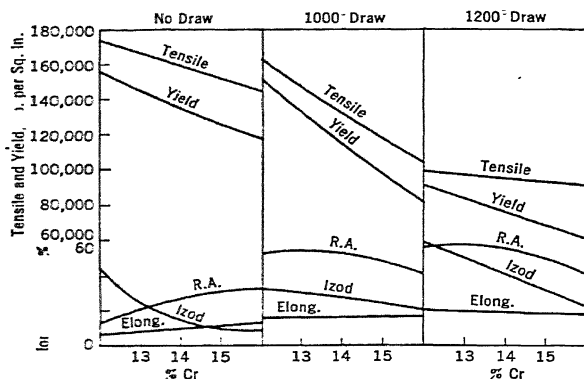


FIG. 17.—Steels of 0.09 C, 12 to 16% Cr, quenched from 1850°, drawn as shown. (Giles.<sup>10</sup>)

See general comment on property charts, p. 40, chap. 1.

ple, architectural trim, and the like, where final mechanical properties are of little moment and workability and corrosion resistance are paramount.

**Sheet and Strip.** In a so-called malleable type, employed among other uses, for sheet and strip, the maximum of 0.12% C may be the same as for the turbine grade, but for formability under 0.10% C may be specified. The Cr content may vary more widely, 11.50 to 16% being included in the several commercial grades. With the higher Cr contents the atmospheric corrosion resistance is good even in the as-rolled condition. If workability is desired along with high corrosion resistance, hardening, followed by drawing at 1200–1400°, is employed. Giles<sup>10</sup> has shown the range of properties in Fig. 17. Corrosion resistance of these steels is good when the steel is hard, or when almost fully softened, but not so good when the steels are drawn in the 600–1200° interval. In the lower C range, at about 12.5% Cr, the sheet

may be sufficiently work hardened by cold rolling to make a cheap grade of stainless cutlery without heat treatment. (The better grades of stainless cutlery, however, are made from steels with 13 to 18% Cr with C ranging from 0.35% in the lower Cr, to 0.55 to 1.10% C in the higher Cr grades. Even the first named composition is hypereutectoid.) Chromium steels of still higher C are used for special purposes.

**Special Bearing Steel.** The upper end of the Cr and C ranges just mentioned are combined in steel for special ball bearings, valve seats, etc., where great wear resistance as well as corrosion resistance must be had. Steel of around 1.00 C, 17.50% Cr has been discussed by Morton and Rummeler,<sup>11</sup> who give detailed instructions for its treatment. Around 0.50% Mo and 0.25% V may appear in the steels in the higher Cr range. These special ball-bearing steels with C at 1% or even considerably more may be used as a "mild high-speed" steel since, when the service does not require too much red hardness, they behave something like a W high-speed steel. It is not of great importance as a tool steel but is of interest as indicating how the properties and uses of the alloy steels high in carbide-forming elements merge and overlap.

**Special Die Steels.** Blanking and drawing dies, and the like, may be made from Cr steels of 1.35–2.35% C and 12–17% Cr. These "high-C, high-Cr" steels may have small additions of Ni, Mo, V, and Co, singly or in combination. Such steels contain many carbide particles and are used for their wear resistance. With so much C, the corrosion resistance is small compared to low-C steels of equal Cr content. The die steels will be discussed in Chap. 22.

**Cutlery Steels.** The cutting power and the ability to hold an edge of all cutlery and fine-edged tools are closely related to hardness of the matrix plus presence of finely divided and well-distributed excess carbides. It happens that in the cutlery grades of stainless, the corrosion resistance is also at a maximum in the hardened or hardened and low-temperature drawn state. They rust quite readily in the annealed state. This is traceable to the C content.

The early type of cutlery stainless contains about 0.35 C, 13% Cr. On oil quenching from 1800–1850° it hardens to 56 Rockwell C. A later type, developed for equal stainlessness and higher hardness, as is desired in surgical instruments, for example, contains about 0.70 C, 16–17% Cr, often with 0.50% Mo. Oil quenched from 1800–1925°, this hardens to about 60 Rockwell C.

These steels are stiff at high temperature and should not be forged below 1700°. On the other hand they are readily burnt, so that heating much above 2000° for forging should be avoided. They have low thermal conductivity, and require slow heating, preferably a preheat-



ing, say at  $1300^{\circ}$  to  $1500^{\circ}$ , analogous to that given a high-speed tool. The steels are sluggish and require high temperatures and long times for homogenization. High quenching temperatures improve the stainlessness but excessive ones involve grave danger of burning. The hardness is but slowly affected by tempering up to  $800^{\circ}$ .

Process annealing at  $1400$ – $1450^{\circ}$  is used to improve machinability, full annealing requires soaking at  $1600^{\circ}$  and slow cooling, in either grade. Stress relief annealing is carried out at around  $1300^{\circ}$ . The properties of the ordinary 0.30% C grade are shown in Fig. 18.

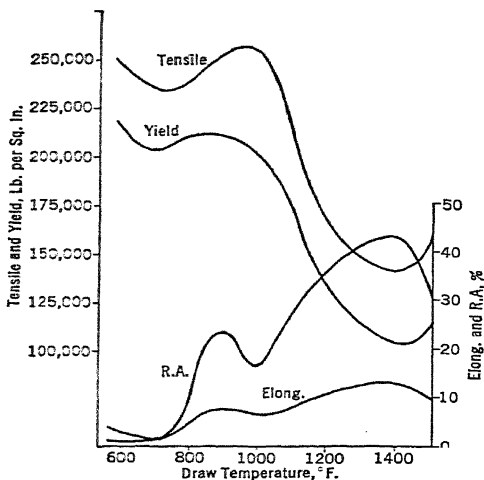


FIG. 18.—Cutlery stainless steel, 0.30 C, 13% Cr. Oil quenched from  $1825^{\circ}$ , drawn as shown. (Parmiter.<sup>25</sup>)

See general comment on property charts, p. 40, chap. 1.

**Other Uses.** A steel of 0.25 C, 14 Cr, 0.75% Si, oil quenched from  $1500^{\circ}$  and drawn at  $1250^{\circ}$ , giving 120,000 tensile, 75,000 yield, 21.5 elong., 60% R.A., was experimentally tested<sup>12</sup> for machine gun barrels, and, once its machining was solved, produced a barrel with the least erosion and greatest accuracy after long use of any steel tried. Some sporting rifle barrels have been commercially produced.

**Machinability.** The machinability of these steels is poor, but by addition of 0.30–0.40% S or Se, they can be made to machine almost like screw stock. The S is sometimes incorporated as zirconium sulphide or molybdenum sulphide. The corrosion resistance is not notably decreased by the additions. The ductility and impact fall with such large additions. An addition of Cu in excess of its solubility, so that

particles of free Cu are present, is effective in increasing machinability.<sup>13</sup>

**Cast Chromium Steels.** Since ductility is a prime requirement in cast steels and Cr tends to reduce it, the plain Cr steels find only limited use in castings. Grotts<sup>14</sup> gives for 0.36 C, 0.74 Mn, 0.41 Si, 0.83% Cr normalized 1650° drawn 1250°; 95,000, 55,000, 18, 31 and 29 Izod. Hence Cr usually enters the cast steels only in conjunction with other elements that improve ductility. The Cr is used in these for depth-hardening properties and to confer wear resistance.

**Five Per Cent Chromium Castings.** Castings of the 5 Cr, 0.50% Mo composition, with C around 0.20–0.30%, are used for return bends, valves, and the like in oil-refinery service. They are slow cooled in the mold, normalized from 1750–1950°, and drawn at 1200–1250°. Double normalizing with the second treatment at 1650–1700° and drawing at 1225° give around 115,000 tensile, 80,000 yield, 18% elong., and 40% R.A.

**Martensitic Castings.** For resistance to wear and seizure, coupled with fair resistance to corrosion, such as is required in some pump liners, valve seats, etc., air-hardening Cr cast steels of around 0.30 C, 12–14% Cr, or of around 0.80 C, 16–19% Cr, each with about 0.40 Si, 0.50 Mn, and 0.50% Ni are employed. These are either normalized (air quenched) or oil quenched from 1850–1875° and drawn at 500–600° to 55 Rockwell C.

**Cast 16–20 Per Cent Chromium Alloys.** In the next higher range of Cr content, 16–20%, the castings may be either entirely ferritic and not amenable to heat treatment, or responsive to heat treatment like the martensitic grade just mentioned. When the composition is held within 0.20–0.30 C and 16–20% Cr, with about 0.75 Si, 0.50 Mn, and up to 1% Ni, the steels are ferritic, and are annealed at 1500–1550° for an extended period and slow cooled. These alloys are strong, 80,000–105,000 tensile, 55,000–75,000 yield, but not tough, elong. 5–12%, R.A. 6–15%. The strength increases as the C, Cr, and Ni contents increase.

By raising the C to 0.70–1.00% the steel becomes hardenable, and is quenched from 1850–1875° and drawn back as desired. The castings of this group are sensitive to pouring temperatures and section size. Hot pouring and slow cooling produce very large crystal size and decreased toughness.

**Twenty-five Per Cent Chromium Castings.** For maximum scale resistance a Cr content of nearly 25% is needed, so castings with Cr of 25–30% are produced. The grain-size problem is accentuated at this Cr level, but the addition of nitrogen (as high N ferrochromium which can be rather readily prepared since the very high Cr alloys

when molten readily take up N from the air) serves to limit the grain size.

At best these cast alloys are not very ductile. An alloy with 0.35–0.45 C, 25% Cr has a tensile of about 60,000–85,000, yield 45,000–55,000, with only a few per cent elong. or R.A. as cast. By annealing at 1500° and slow cooling to 1100°, then quenching, the ductility may be slightly increased. Such treatment also improves machinability. The N-containing alloys with around 0.20% N may show over 100,000 tensile, 70,000 yield, and over 10% elong. and R.A., but careful foundry practice is needed to cast the high-N alloys entirely sound.

When wear resistance is sought, the C content may be increased, as high as 3% C sometimes being used, with 0.75% C common when resistance to abrasion and to high temperature corrosion by SO<sub>2</sub> or S vapor are to be combined.

**Wrought Alloys.** Wrought Cr-Fe, in the range of about 16–30% Cr, with lower C than is customary in castings, e.g., 0.10–0.15% C, is susceptible to deterioration at 750–1050°. Although tough at such temperatures, when the steel has been held for a long time in this range and is then cooled to room temperature, it may be very brittle. Heating to 1300–1400° will restore the toughness. This is evidently a carbide precipitation-hardening effect. The sensitive temperature zone accounts for the quenching from 1100° in the treatment used for toughening castings. Corrosion resistance of these alloys is not much changed even when embrittlement has occurred.

In the lower C wrought materials of this class the carbide solution and precipitation may be avoided by addition of sufficient Ti or Cb, just as the martensitic hardening of the 5% Cr steel can be similarly avoided. The embrittlement and its avoidance by binding the C with some stronger carbide former than Cr are analogous to the behavior of high Cr austenitic steels, and will be discussed in Chap. 20.

Tofaute<sup>22</sup> suggests, as substitute for the more expensive 18 Cr 5% Ni, a steel of 0.08 C, 18 Cr, 2% Mo, with or without a Ti addition. This is said to be especially suitable for use against organic acids, and to be as workable in deep drawing as 18 : 8.

**Summary.** Chromium is primarily a carbide former and thus tends to go into carbide, though in low-C steels, or after the C has been satisfied, it acts as a ferrite strengthener, in a manner analogous to Mn. Its tendency toward carbide formation, however, is much more marked. It is a gamma-loop-forming element. Ac<sub>3</sub> is materially raised by large amounts of Cr, but in the lower percentages, when accompanied by C, it may be lowered, the position of the curve being

strongly influenced by the C. The eutectoid C content is materially lowered by Cr.

The Cr-bearing carbides are very hard, hence are used to confer wear resistance and cutting ability. They are quite stable at high temperature and impede graphitization. The carbide-forming tendency finds use in carburizing steels. The carbides are thought to go into solution in austenite with difficulty, thus acting as inhibitors to grain growth. When sufficient Cr is dissolved in austenite it is made sluggish, its critical cooling rate decreased, and depth hardening produced. Tempering of martensite and troostite is somewhat retarded. Chromium steels may have a tendency toward temper brittleness.<sup>1</sup>

Just as the Mn : C ratio is important in Mn steels, so the Cr : C ratio is important. If the C is too high there is too little ferrite strengthening and the steels tend toward brittleness. On the whole the Cr steels tend to have, for a given strength somewhat lower ductility and a higher yield ratio than some of the other alloy steels. Chromium is essentially a hardener rather than a toughener, hence is very often accompanied by other elements that confer toughness.

Chromium steels are usable over a very wide range of Cr and C contents, the high-C series comprising tool and cutlery steels, the low-C steels providing carburizing steels. These have low-Cr content; those useful for high temperature oxidation resistance have intermediate to high Cr; those for corrosion resistance and "stainlessness" have high Cr. The corrosion resistance is a highly specific property.

Austenitic steels of high-Cr content and complex steels with Cr are of great importance and will be dealt with in Chaps. 20 and 13, respectively.

There are books covering the properties of Cr steels in detail by Monypenny,<sup>15</sup> Greaves,<sup>7</sup> Thum,<sup>16</sup> and the first volume<sup>1</sup> of a two-volume monograph, "The Alloys of Iron and Chromium," dealing with up to 10% Cr. A second, on the high Cr alloys, is in preparation by the Alloys of Iron Committee. Complex Cr steels are also dealt with in other volumes of the Alloys of Iron Monograph Series.

#### BIBLIOGRAPHY

1. KINZEL, A. B., and W. CRAFTS: "The Alloys of Iron and Chromium." New York, McGraw-Hill Book Co., Vol. 1, 1937, 535 pages.
2. WRIGHT, E. C., and P. F. MUMMA: Properties of Low-Carbon Medium-Chromium Steels of the Air-Hardening Type, *Trans. Am. Inst. Min. Met. Eng.*, v. 105, 1933, pp. 77-87.
3. ESSER, H., and H. MAJERT: Einfluss der Abschrecktemperatur auf die Stahlhärtungsvorgänge (Effect of Quenching Temperature on the Hardening of Steel), *Archiv f. Eisenhüttenw.*, v. 7, 1933, pp. 319-322.

4. BAIN, E. C.: Some Characteristics Common to Carbon and Alloy Steels, *Yearbook Am. Iron Steel Inst.*, v. 24, 1934, pp. 86-128.
5. AITCHISON, I.: Chromium Steels and Irons, *Proc. Inst. Automobile Eng.*, v. 16, 1921-1922, part 1, pp. 183-217. Also: *Engineering*, v. 112, 1921, pp. 771-772, 805-807.
6. JONES, J. A.: The Properties of Some Chromium Steels, Research Department, Woolwich, Report 51, 1921, 19 pages.
7. GRAVES, R. H.: "Chromium Steels." London, Department of Scientific and Industrial Research, His Majesty's Stationery Office, 1935, 321 pages.
8. FLEISCHMANN, M., and S. D. WILLIAMS: Aluminum Alloy Steels for Sour Crude Refining, *Natl. Petroleum News*, v. 29, July 28, 1937, pp. R51-55.
9. MOCHEL, N. L.: Low Carbon Stainless Steels; Forgings and Rolled Bars (in E. E. Thum: "The Book of Stainless Steels," 2nd Ed., pp. 251-266).
10. GILES, D. J.: Sheet and Strip, *Ibid.*, pp. 267-276.
11. MORTON, H. T., and I. A. RUMMLER: Forging Steels Containing 1% Carbon, *Ibid.*, pp. 292-297.
12. DE SVESHNIKOFF, V. V.: Data on Some Special Steels for Machine Gun Barrels, *Trans. Am. Soc. Metals*, v. 21, 1933, pp. 652-662.
13. LIPPERT, T. W.: 10 Cu-18 Cr-72 Fe, etc.—Digby's Cypritic Alloy, an Entirely New Species of the Genus Stainless Steel, *Iron Age*, v. 140, October 14, 1937, pp. 54-63.
14. GROOTS, F.: Representative Properties of Cast Medium Pearlitic Steels, A.F.A.-A.S.T.M. "Symposium on Steel Castings," 1932, pp. 159-174.
15. MENYENNY, J. H. G.: Chromium-Manganese Steels as Heat-Resisting Steels, *Metallogra.*, v. 17, 1938, pp. 93-96.
16. THUM, E. E. (editor): "The Book of Stainless Steels." Cleveland, The American Society for Metals, 2nd Ed., 1935, 787 pages.
17. TOEFAUTE, W.: Über sparsstoffarme, nichtrostende und säurebeständige Stähle (Inexpensive Rust Resistant and Acid Resistant Steels), *Tech. Mitt. Krupp*, v. 6, April, 1938, pp. 17-24.
18. EPSTEIN, S., J. H. NEAD, and J. W. HALLEY: Choosing a Composition for Low-Alloy High-Strength Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 120, 1936, pp. 309-345. Also: *Metals Technology*, v. 3, 1936, Tech. Publ. 696, 31 pages.
19. MAURER, E., and H. NIENHAUS: Über den inneren Aufbau der Chromstähle (The Inner Structure of Chromium Steels), *Stahl u. Eisen*, v. 48, 1928, pp. 9
20. Bethlehem Steel Company: "Bethlehem Alloy and Special Steels." Bethlehem, Pa., The Author, 1935, 375 pages.
21. KRIVOBOK, V. N., and M. A. GROSSMANN: A Study of the Iron-chromium-carbon Constitutional Diagram, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 760-807.
22. BAIN, E. C.: X-rays and the Constituents of Stainless Steel, *Ibid.*, v. 14, 1928, pp. 270-275.
23. WRIGHT, E. C.: Five Per Cent Chromium Steels: Rolled Material, Especially Tubing (In E. E. Thum: "The Book of Stainless Steels," 2nd Ed., pp. 234-244).
24. MACQUIGG, C. E.: 25 to 30 Per Cent Chromium-iron Alloys, *Ibid.*, pp. 351-368.
25. PARMITER, O. K.: High Carbon, High Chromium Stainless Steels, *Ibid.*, pp. 278-292.

## CHAPTER 8

### THE ASSISTANT ELEMENTS, ALUMINUM, TITANIUM, ZIRCONIUM, VANADIUM, ETC.

Although we set out to deal with the simple alloy steels in the early chapters of this volume, the discussion of every alloying element has had to include reference to its being bolstered up by small amounts of other elements, in order to impart to the steel some desirable attribute not sufficiently provided by the single element. By this time, the reader should be impressed by the fact that an alloy steel with only one alloying element is quite a rarity.

Most of the alloying elements heretofore discussed are (with the exception of Mo, at about 0.2%) used in constructional alloy steels in amounts running from, say 0.50%, up. The effect of smaller amounts is seldom outstanding, and 1% or more is often needed before the effect is really strong. Although even these elements are, more often than not, accompanied by other alloying elements to augment or to correct the tendencies conferred by the main elements, each of them does find sufficient use by itself, with C, to have made it necessary to consider them individually.

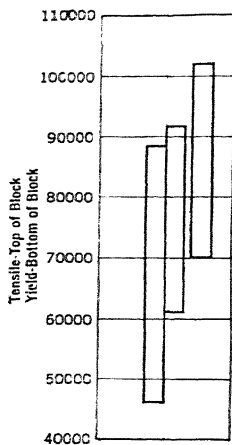
There is another group of alloying elements which exert strong effects but which are normally utilized only in small amounts, rarely alone, almost invariably in conjunction with other elements. They may be utilized because of their effect upon grain size, because of their influence on depth hardening, or their ability to enter into complex carbides, to harden these carbides and to stabilize them for high temperature use. To a subsidiary degree they are employed for their ferrite-strengthening properties.

**Aluminum.** Prominent among the elements used to control the propensity toward grain growth is Al. In Chap. 21 the effect of 0.01% Al on C tool steel is discussed, and the general topic of grain growth and of its control by Al has been dealt with in Chap. 9, Vol. I.

Steel that has been made resistant to strain aging and quench aging by addition of Al has long been in use in Germany under the name "Izett" (from "immer zäh," "always tough"), especially in low-C, plain C grades for boiler steel, because of the belief, expressed

# ASSISTANT ELEMENTS. ALUMINUM, TITANIUM, ETC.

by Neuendorff,<sup>1</sup> that it was resistant to the so-called "caustic embrittlement," a subject that, as Schroeder, Berk, and O'Brien<sup>2</sup> have shown, may not be so simply solved. However, the toughness of Izett is clearly shown by Neuendorff and is also indicated in Fig. 23, Chap. 1, Vol. I. As is discussed in Chap. 9, Vol. I, Herty and co-workers,



The middle block is for double normalized V steel, the right-hand one for single normalized.

See text, p. 162, for details of treatment.

See general comment on property charts, p. 40, chap. 1.

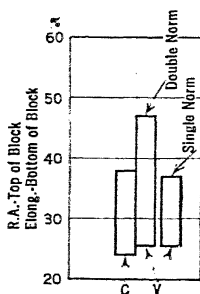


FIG. 1.—Properties of 0.30% C cast steels, with and without 0.20% V.

as well as many other investigators, have shown that the toughness is primarily a phenomenon connected with the retention of small grain size, irrespective of the means by which fine grain is preserved. It is this fact that justifies lumping Al, Ti, Zr, and V together as elements useful for grain-size control.

It does not suffice to dump a little Al into a heat of steel in order to produce a good fine-grained i.e., grain-growth-resistant steel; the melt must be in proper condition to receive it. As Epstein, Nead and Washburn<sup>3</sup> have shown, the production of fine-grained steel by the use of Al is now pretty well understood. Making a steel of intermediate grain size, i.e., of medium propensity toward coarsening, is more difficult. Sugar<sup>23</sup> also discusses methods of adding Al for grain size control.

Amounts of Al of the order of 1% or more, oddly enough, tend to result in coarse-grained steels, and large additions are made only in very special steels, such as those for nitriding (Chap. 15) or in certain heat-resistant compositions (Chap. 19), though Foley<sup>4</sup> has described a precipitation hardenable Ni-Al steel with interesting propensities (see Chap. 8, Vol. I).

**Aluminum Additions to Cast Steel.** Although the addition of Al to cast steel aids in controlling grain growth, as in wrought steels, its primary use has been to ensure sound, gas-free castings. Such an addition is very prone to lower the ductility in acid electric steel. In discussion of a paper by McCrae and Dowdell<sup>5</sup> Lorenz cites data on 5 steels, ranging from 0.35 C, 0.80% Mn to 0.25 C, 1.50% Mn, with and without Al addition, tested annealed, normalized and quenched and drawn. Lumping the results of all treatments, the average was

	TENSILE	YIELD	ELONG.	R.A.	CHARPY
No Al.....	81,000	50,000	27.5	47	20.5
Trace Al added.....	81,000	51,000	23	31.5	13

The production of cast steel that is both free from gas holes and of high ductility would be much facilitated if the proper utilization of Al could be more certainly worked out. Sims and Lillieqvist<sup>6</sup> find that Al addition to a steel that has already been freed from almost all its FeO by other deoxidizers is almost certain to produce low ductility, whereas if the steel is well oxidized before addition of Al, the ductility is seldom injured. They point out that poor ductility appears to be the result of grain boundary inclusions of a eutectic-like appearance, and that globular inclusions do not lead to low ductility. In later work by Sims and Dahle<sup>7</sup> it appears that the S content may be an important factor, and that it is quite likely that the amount of Al needed bears a relation, not only to the FeO content but also to the S content. The indications are that a more generous addition of Al than has been customary in steel foundries may avoid the difficulty, ensure freedom from porosity, good ductility, and the necessary resistance to grain growth.



When an unsuitable addition of Al has been made, its effect is less harmful if the C content is low. McCrae and Dowdell found that Al-treated steels of 0.17% C met the specifications not only for ductility but also for tensile and yield strengths. In higher C steels they found indications that a high normalizing temperature, 2000°, may help in raising the ductility of Al-treated cast steel that would show too low ductility if normalized from 1650°, but experiments were not carried far enough to substantiate this impression.

**Titanium, Zirconium, Columbium.** The similarity of the behavior in normalizing of cast steels with addition of about 0.05–0.20% of either Al, Ti, or V, instead of the very small amounts of the former two used for deoxidation alone, is evident when the data for Mn-V and Mn-Ti steels (Chap. 3) and for Cu steels with V, Ti, or Al (Chap. 5) are studied. Duma<sup>5</sup> states that the Norfolk Navy Yard uses 0.05–0.20% Ti in all its steel castings in order to secure the desired combination of strength and ductility, such an addition being made even to a Ni-Cr-Mo cast steel. However, in the cases he cites, where some oxide of Ti is added to the bath, plus Al to reduce it, some of the Al goes into the steel, so the results are not necessarily due to Ti alone. He finds more than 0.20% Ti to be detrimental.

In general, Ti or Zr may be used, like Al, for grain-size control in wrought steels and for producing soundness in steel castings. Like Al, these elements combine with O, N, and S. Unlike Al, Ti, at least, very readily combines with C in steel to form carbide—its behavior in this respect being especially evident in the plain Cr and the 18–8 stainless steels (see Chaps. 7 and 20), where it, as well as Cb, finds use as a stabilizer against carbide precipitation.

Titanium and Cb find use as an "antidote" for C in various steels in which undesired carbide precipitation may take place. Such carbide precipitation is met in both ferritic and austenitic Cr steels (see Chaps. 7 and 20), and either the cost of low-C ferrochromium compared to high-C or the necessities of manufacturing methods may preclude holding the C content low enough to prevent it. In such cases the addition of sufficiently large amounts of Ti or Cb in proportion to the C content appears to bind the C as a stable carbide and make the ferrite or austenite of the matrix act as if the alloy were C-free, ferritic steels so treated being non-hardenable.

Recent information on Ti includes an account, by Swinden and Reeve,<sup>24</sup> of its use in improving the suitability of low alloy structural steels containing Mn for welding, and a study of the Fe-Ti-C system by Tofaute and Büttlinghaus.<sup>25</sup> The latter shows that in carbonless Fe the gamma loop extends to 0.80% Ti. In the presence of C, and

0.30% Ti, the eutectoid moves back to 0.50% C and  $Ac_1$  is raised to 1350°. At 0.60% Ti, the eutectoid is at 0.35% C and  $Ac_1$  at 1730°.

Comstock and Clark<sup>9</sup> have discussed the effect of Ti in a plain 17.5% Cr steel.

Not much is said in the literature about Zr additions for grain-size control, although such additions are made in commercial practice. The effect of Zr in improving impact resistance, noted long ago,<sup>10</sup> is doubtless due to its effect in reducing grain-coarsening propensities, though this was not clearly brought out in the early work.

More than one of the deoxidizing and grain-controlling elements may be used. Both Al and Ti may be used in making "non-aging" steels (see Chap. 9, Vol. I). Complex ferroalloys used as deoxidizers are on the market (some of which are intentionally produced as primary products, others are by-products of certain metallurgical processes), in which Al, Si, and Ti, or Si, Al, V, and Zr are present. Some grades of ferroalloys of Ti, Zr or V made by alumino-thermic reduction methods contain Al; likewise sometimes those made by electro-thermic methods contain Al coming from reduction of Al compounds in the ore. Ordinary ferrosilicon contains 1 to 2% Al and around 0.10% Ti, along with its 50% Si. Since very small amounts of the grain-controlling elements may be effective, if ferroalloy additions containing them are made after the heat has been so thoroughly deoxidized by previous additions that the elements are not completely oxidized and slagged off, the effect of the ferroalloy may be influenced by the presence of the elements, so that different grades of ferro may behave differently, by virtue of the variable amounts of Al or Ti thus added.

**Other Elements.** Krivobok<sup>11</sup> has hinted that boron may find a place in stainless steels; Herzig<sup>12</sup> suggests it in Mo high-speed steels; and in conjunction with W, Mo, and Cr, it already finds use in welded-on "overlays," for wear resistance. The hardening and grain-refining effects of 0.01 to 0.10% boron in steel has been discussed by Naton and Piwowarsky.<sup>13</sup> Tantalum has been suggested as an alloying element in steels for nitriding and is also claimed<sup>14</sup> to reduce grain size in other steels. Uranium and cerium<sup>15</sup> have been studied to some extent, but, with the melting technique in use when they were tried, both were so prone to produce dirty steel that their possibilities have not been recently explored. Thus, future development may bring into use, either as alloying or as "assistant" elements, several that are not now listed as such.

**Vanadium Steels.** A small V addition, often 0.08 to 0.12% and normally less than 0.20%, is often made to plain C and to a variety of

alloy steels, especially to the Cr steels. The Cr-V steels will be discussed in Chap. 13.

authorities, notably Aitchison,<sup>16</sup> have strongly questioned the properties of the Cr steels, in particular, are enhanced by the V addition; others have questioned the economics of such an addition, though admitting the technical effect. The situation has been clarified by recent advances in knowledge of grain-size control in general, and in particular by the work of Zimmerman, Aborn, and Bain.<sup>17</sup>

Vanadium is often used in amounts of 0.05–0.10% or even less, in steels in which its presence is not specified, being added purely for certainty of grain-size control and prevention of low coarsening temperatures. Advocates of V claim that it is easier to get uniformity in coarsening characteristics from heat to heat by the use of V than by Al.

**Effect in Normalizing.** An outstanding effect of V is in improving the properties of normalized steels. In a C steel normalized from 1520° and a C-V steel from 1565°, both drawn at 1000°, the following properties are obtained:

% C	% Mn	% Si	% V	TENSILE	YIELD	ELONG. R.A.		Izod
						%	%	
0.50	0.71	0.19	...	103,000	60,500	26	48	18
0.50	0.76	0.37	0.27	110,000	80,500	26	54.5	42

The large jump in yield strength and in impact is characteristic. Similar effects are produced by smaller doses of V.

Vanadium forms a gamma loop and raises  $Ac_3$ , but in the small amounts ordinarily used in other than tool steels the effect is inappreciable.

**Cast Vanadium Steel.** The average values of Fig. 1, Chap. 1, for plain 0.30% C normalized cast steel are compared in Fig. 1 with data from Barton<sup>18</sup> for a 0.30 C, 0.75 Mn, 0.35 Si, 0.20% V cast steel double normalized from 1700° and 1400° and single normalized from 1700°. The effect of the V in increasing both strength and ductility on double normalizing, and strength without effect on ductility on single normalizing, is evident.

**Wrought Vanadium Steels.** The same effect is shown in quenched and tempered steels. Figure 2 compares data, including impact, on wrought water-quenched and tempered C and C-V steels. The C-V steel is higher in C, so the ductility of both steels is of the same order; but the tensile and yield strengths are markedly higher in the C-V. Even with the higher tensile, the C-V steel has a decided advantage in impact.

The effects of V in allowing higher temperature of heating for quenching, thus a high quenching temperature and resultant increase in tensile and yield strength, and in retaining fine grain and resultant high ductility and impact resistance, are brought out in Chap. 6 for Ni-V, in Chap. 13 for Cr-V, and for Mn-V cast steels in Chap. 3. All these cases combine the use of a depth-hardening element plus a grain-refining element. Quite analogous is the "Amola" steel discussed in Chap. 9, in which Mo adds depth hardening and Al is used for grain control.

**Effect of Vanadium Carbide.** A small "pinch" of V, say between 0.05 and 0.20%, suffices, according to the usual explanation, to produce V-rich carbides which are with difficulty soluble in austenite. Until these carbides are taken into solution the steel does not coarsen materially. As the temperature is raised and the solution of the carbides approaches completion the grain size increases slowly in the temperature range, where, in the absence of V, coarsening occurs with a jump. However, when the temperature is so high that the last traces of the V-rich carbide are absorbed, then grain growth does occur, and the grain may even be coarser than in a similar steel, without V. This jump, in the V-steels, however, occurs at temperatures so far above the normal heating range that in regular practice the V-containing steels remain fine grained at normal heating temperatures, hence are tough, and rather shallow hardening. When a steel is to be treated by normalizing rather than quenching and tempering, the maintenance of toughness, in spite of what would be overheating without V, allows much more leeway in temperature. For quenching, there is the possibility of heating to such a temperature that only slight grain growth occurs, thus retaining toughness but yet securing deeper hardening than in a comparable steel that insists on staying fine grained. The V steels appear to be more suited to such a compromise treatment than those made fine grained by use of Al.

However the addition of V to a C steel is primarily to aid it to retain fine grain and toughness. It can confer only a minor degree of depth hardening, so V is generally combined with a depth-hardening element, most commonly the cheap ones, Mn or Cr.

**Manganese in Vanadium Steels.** Thus what are commonly called C-V steels are almost invariably really Mn-V steels, so that even there the V actually plays the rôle of an auxiliary element.

**Temperature Effects.** The V-rich carbides not only appear to dissolve more slowly in austenite but to be more resistant to high temperature and less soluble in ferrite than is cementite, so the V steels show resistance to tempering, especially in the range 850–1200°. This

tends to confer stability and improved strength at high temperatures upon pearlitic steels containing V.

**Special Vanadium Steels.** The stability of the V carbides, together with the retention of toughness due to their presence, leads to the

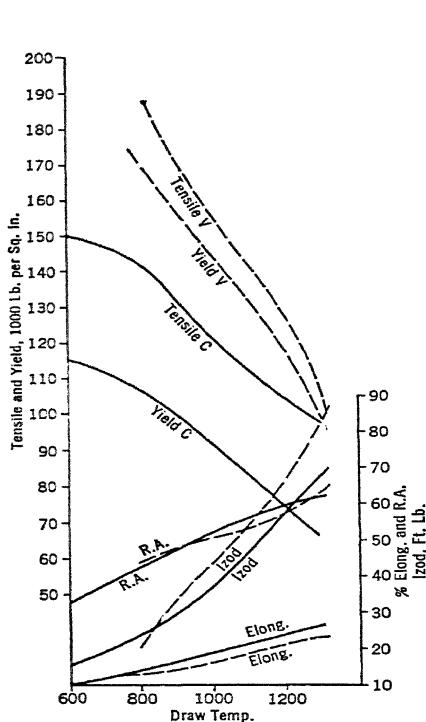


FIG. 2.—Properties of wrought steels. Solid lines, 0.45 C, 0.65% Mn, No V, 1500°, water. Dashed lines, 0.49 C, 0.77 Mn, 0.16% V, 1550°, water. Drawn as shown. (Bethlehem Alloy Steels, Vanadium Steels and Irons.<sup>21</sup>)

See general comment on property charts, p. 40, chap. 1.

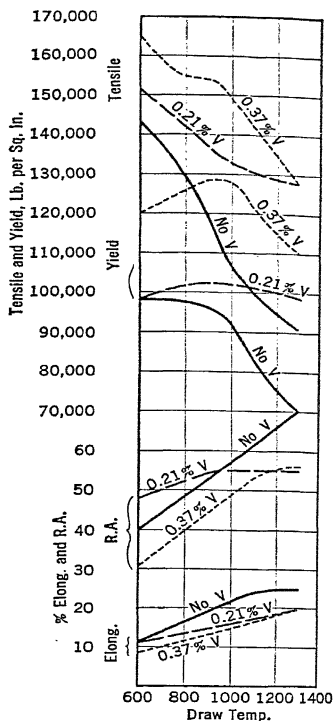
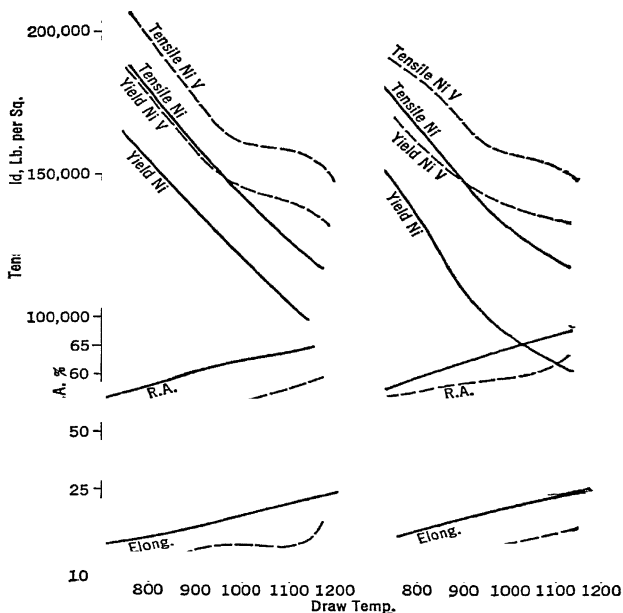


FIG. 3.—Effect of V in steels of 0.36–0.39 C, 0.32–0.39 Si, 0.45–0.59% Mn with varying V, oil quenched and drawn as shown. (After Houdremont, Bennek and Schrader.<sup>19</sup>)

introduction of V into practically all grades of tool and die steels, especially for battering tools where toughness is paramount. In such cases, where the C content must be high to supply free carbides, the V

content rises correspondingly above the approximately 0.10% that serves when its rôle is only for grain-size control.

Spring steels offer an important field for use of V since it confers certainty of fine grain and toughness, even at high hardness. There-



Water quenched. Oil quenched.  
Solid lines, 2330, 0.30 C, 0.65 Mn, 3.50% Ni, quenched from 1450°.  
Dashed lines, 0.29 C, 0.45 Mn, 3.41 Ni, 0.20% V, quenched from 1600°, then from 1520°.  
(Nickel Alloys Steels and Bullens.)

FIG. 4.—Comparison of Ni and Ni-V steels.

See general comment on property charts, p. 40, chap. 1.

fore Cr-V and Si-V spring steels are widely employed. (See Chap. 16.)

**Precipitation Hardening.** Houdremont, Bennek, and Schrader<sup>19</sup> present data that are plotted in Fig. 3 for steels of 0.36–0.39% C, 0.32–0.39% Si, 0.46–0.59% Mn with no V, 0.21% and 0.37% V, quenched in oil and tempered as shown. A precipitation-hardening effect, ascribed to separation of a V carbide at 900–1000° is noted. Such behavior, a “secondary hardening,” is met in various tool steels containing V.

**Nickel-Vanadium Steels.** The data tabulated in Chap. 6 for a Ni-V steel containing 0.20% V are plotted in Fig. 4 in comparison with a similar steel free from V. The results for the two quenching temperatures used on the Ni-V steel have been arranged in the plot. Several factors combine to make the Ni-V steel stronger and less ductile at a given draw than the plain Ni steel, the higher quenching temperature, a strengthening due to the relatively large V addition, and the resistance to tempering when cementite contains V carbide.

The tendency toward a secondary hardening effect, noted in Fig. 3, is detectable in Fig. 4. The same tendency will be noted in Fig. 4, Chap. 12.

The curves indicate that V has an alloying effect as well as a grain-refining effect, though the latter predominates when the V content is held down to the usual 0.10–0.12%.

The V steels have been described by Hougardy<sup>20</sup> and by the Vanadium Corporation of America.<sup>21</sup> Since the elements discussed in the chapter are "assistant elements," their effects are mentioned, in one way or another, in nearly every chapter of this volume.

#### BIBLIOGRAPHY

1. NEUMENOFF, G. G.: Izett Steel. *Metals & Alloys*, v. 3, 1932, pp. 61–68.
2. SCHNEIDER, W. C., A. A. BERK, and R. A. O'BRIEN: Intercrystalline Cracking of Steel in Aqueous Solution, *Ibid.*, v. 8, 1937, pp. 320–330.
3. EISEN, S., J. H. NEAD, and T. S. WASHBURN: Grain-size Control of Open-Car Steel. *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 942–978.
4. F. B. Nickel-Aluminum Compound—A Hardener for Steel. *Metal Progress*, 29, May, 1936, pp. 53–56.
5. McCRAE, J. V., and R. L. DOWDELL: Effects of Deoxidation and Mold Conditions on the Tensile Properties of Carbon Steel Castings, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 159–211.
6. SIMS, C. E., and G. A. LILLIEQVIST: Inclusions—Their Effect, Solubility, and Control in Cast Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 100, 1932, pp. 154–195.
7. SIMS, C. E., and F. B. DAHLE: Effect of Aluminum on the Properties of Medium Carbon Cast Steel, *Trans. Am. Foundrymen's Assoc.*, v. 46, September 1938, pp. 65–132.
8. DUMA, J. A.: Effect of Titanium on Some Cast Ferrous and Nonferrous Metals, *Trans. Am. Soc. Metals*, v. 25, 1937, pp. 788–825.
9. COMSTOCK, G. F., and C. L. CLARK: Effect of Titanium on Some Properties of 17.5% Chromium Steel, *Metals & Alloys*, v. 8, 1937, pp. 42–46.
10. FEILD, A. L.: Some Effects of Zirconium in Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 69, 1923, pp. 848–894.
11. KRIVOBOK, V. N.: A Digest of Stainless Steels, *Yearbook Am. Iron Steel Inst.*, v. 27, 1937, pp. 129–167.
12. HERZIG, A. J.: High Speed Alloy Steel, U. S. Patent 2,109,746, March 1, 1938. Also: BREELEER, W. R.: Development in Molybdenum High Speed Cutting Is, *Am. Soc. Metals*, preprint 11, October, 1938, 22 pages.

13. NATON, G., and E. PIWOWARSKY: Einfluss von Borzusätzen bis 0.1% auf die Festigkeitseigenschaften von Stahlguss mit 0 bis 5% Ni (Influence of 0.1% Boron Additions on the Properties of Steel Castings with 0 to 5% Ni), *Archiv f. Eisenhüttenw.*, v. 11, 1937, pp. 283-286.
14. GRECHKO, N., and E. PERELMAN: Influence of Tantalum on the Properties of Steel, *Metallurg.*, v. 11, December, 1936, pp. 64-72.
15. GILLET, H. W., and E. L. MACK: "Molybdenum, Cerium, and Related Alloy Steels." New York, Chemical Catalog Co., 1925, 295 pages.
16. ARCHISON, L.: Chromium Steels and Irons, *Proc. Inst. Automobile Eng.*, v. 16, 1921-1922, part 1, pp. 183-217. Also: *Engineering*, v. 112, 1921, pp. 771-772. 805-807.
17. ZIMMERMAN, J. G., R. H. ABORN, and E. C. BAIN: Some Effects of Small Additions of Vanadium to Eutectoid Steel, *Trans. Am. Soc. Metals*, v. 25, 1937, pp. 755-787.
18. BARTON, L. J.: "Refining Metals Electrolytically." Cleveland, Penton Publishing Company, 1926, 414 pages.
19. HOUDREMONT, E., H. BENNEK, and H. SCHRADER: Härtbarkeit und Anlassbeständigkeit von Stählen mit schwerlöslichen Sonderkarbiden (Hardening Capacity and Resistance to Tempering of Steels with Difficultly Soluble Special Carbides), *Archiv f. Eisenhüttenw.*, v. 6, 1932, pp. 24-34. Also: BURTON, H. H., and T. F. RUSSELL: Precipitation-hardening in Three Steels Containing Vanadium, *Iron Steel Inst.*, Preprint 4, October, 1938, 18 pages.
20. HORGARDY, H.: "Die Vanadinstähle." Berlin, P.u.G. Gärtner, 1934, 224 pages.
21. Vanadium Corporation of America: "Vanadium Steels and Irons." New York, The Author, 1937, 189 pages.
22. Bethlehem Steel Company: "Bethlehem Alloy and Special Steels." Bethlehem, Pa., The Author, 1935, 375 pages.
23. SUGAR, A.: Metallurgical Aluminium, *Blast Furnace Steel Plant*, v. 26, 1938, pp. 406, 408-410.
24. SWINDEN, T., and L. REEVE: Metallurgical Aspects of the Welding of Low Alloy Structural Steel, *Trans. Inst'n Welding*, v. 1, 1938, pp. 7-24.
25. TOFAUTE, W., and A. BÜTTINGHAUS: Die Eisenecke des Systems Eisen-Titan-Kohlenstoff (The Iron Corner of the System Iron-Titanium-Carbon), *Archiv f. Eisenhüttenw.*, v. 12, July, 1938, pp. 33-37. Also: Büttlinghaus, A.: Das System Fe, Fe<sub>3</sub>Ti, TiC, Fe<sub>3</sub>C. Dissertation. *Tech. Hochschule, Aachen*, 1937, 12 pages.



## CHAPTER 9

### PLAIN MOLYBDENUM AND TUNGSTEN STEELS

**Costs, Behavior in Melting.** Both Mo and W are expensive; Mo costs \$0.85 cents per lb., W, at the time of writing, about \$1.80. For most alloying purposes Mo is about twice as potent as W, so that, unless the elements show a marked difference in the properties conferred for a given use, Mo is normally chosen. In general the behavior of the two is so similar that they are best considered together.

Both elements are less oxidizable than Fe and may be reduced into a basic steel melt from the molybdate or tungstate. Hence they join Ni and Cu in being recoverable from scrap and in not complicating steel-making practice.

Molybdenum finds wide use in constructional steels. A Ni-Cr-W carburizing steel has some use in Germany, but in this country entrance of W into any but special-purpose steels is rare.

**Effect on Ferrite.** The ferrite-strengthening effect of Mo in low-C steel is shown in Fig. 1. Percentage for percentage, the effect is of the same order as that of Cu, Cr or Si, less than that of Mn, and more than that of Ni. The marked effect of the addition of fractional percentages of Mo must therefore be more closely related to its effect on the carbides than on the ferrite.

**Effect on Heat-treatment Temperatures.** Molybdenum and W, like Cr and V, are gamma-loop-forming elements, and raise  $Ac_3$ . Figure 2 shows the general position of the gamma loop. In constructional steels, with only small amounts of Mo,  $Ac_3$  is slightly raised. Normalizing and quenching temperatures are, however, usually higher than in analogous steels without Mo because of the behavior of the carbide.

Both Mo and W readily form carbides. These carbides, like those of Cr and V, are difficultly soluble in austenite and ordinarily tend to restrain grain growth. Like Cr, the presence of large amounts of C with Mo or W, leads to the formation of double carbides. This is shown in the Guillet diagram (Fig. 3).

The rise in  $Ac_3$  and the slow solubility of the carbides both allow, and call for, higher hardening temperature than in analogous C steels. The effect of these elements in making austenite sluggish to transform,

is enhanced when the carbides are taken into complete solution, and the solid solution fully homogenized. Thus they show the "T max."

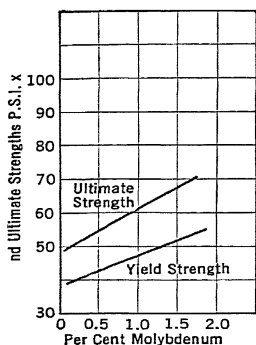


FIG. 1.—Effect of Mo in normalized low C steels. (Epstein, Nead and Halley.<sup>6</sup>)

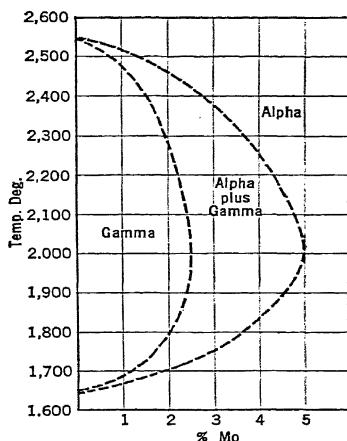


FIG. 2.—The gamma loop in FeMo alloys.<sup>7</sup>

effect (Chap. 6, Vol. I), and the effect upon the position of  $Ar'$  and  $Ar''$  depends on several factors.

**Depth Hardening Effect.** When Mo or W is dissolved in austenite, the austenite becomes very sluggish; thus the depth-hardening and

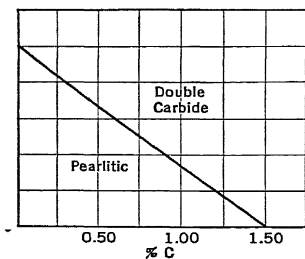


FIG. 3.—Structure of normalized Mo steels. (Guillet.<sup>8</sup>)

air-hardening propensities are very marked, especially with Mo, which, weight for weight, is considerably the more potent of the two. The air-hardening propensity is so great with moderately high Mo content that Mo steels do not normalize to produce the usual type of fine pearlite; instead the structure tends to be "pseudo-martensitic." Neither Mo nor W can be considered as the equivalent of V in suit-

ability for steels to be normalized to pearlite. However, the Mo-containing cast steels are satisfactorily treated by normalizing, provided they are drawn a forward and thin-walled Cr-Mo tubing for aircraft to cool after welding, to secure a degree of air hardening with good results; but these are special cases. In general, the conditions of treatment of Mo-containing steels to which Mo is added are given full heat treatment by quenching and tempering. Constructional Mo-containing steels are almost always complex rather than of the simple C-Mo composition, Mo being used as an "assistant element" to enhance depth hardening.

Both Mo and W carbides, especially their double carbides, are stable at elevated temperatures and for this reason Mo and W are the basic alloying elements in high-speed steels and almost invariably are used in hot-work die steels and the like, as will be discussed in Chaps. 19, 22, and 23.

This stability is evidenced on tempering the quenched Mo or W steels, which require higher temperatures to soften to a given degree

than do C steel. This behavior has been noted for V, but for the amounts normally used is far more marked with Mo and W. In complex steels that are prone to temper brittleness the addition of a little Mo remedies the difficulty. This will be brought out in Chaps. 12 and 13.

#### High Temperature Strength.

Analogous to the reluctance to temper is the general retention of strength at high temperatures. Tungsten behaves, in this regard, much like Mo, but more W is needed and it is more costly, so that, unless other factors come in (such as the way the steels scale on forging) rather than the load-carrying ability at high

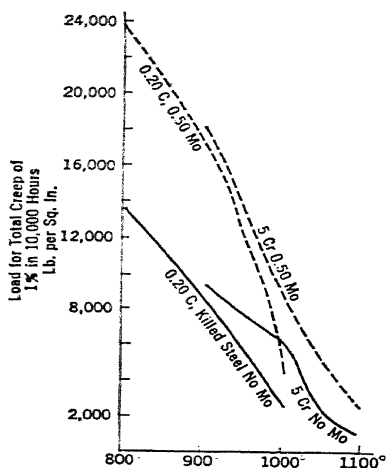


Fig. 4.—Effect of Mo on creep resistance.

temperatures in relation to the alloy cost, Mo is the addition universally chosen to confer high-temperature strength on pearlitic steels. The A.S.M.E Boiler Code Committee<sup>1</sup> sets permissible stresses for C-Mo steel of 0.50% Mo at 950° and 1000° which are more than double those for killed C steel without Mo. Figure 4 shows creep

data from various sources for 0.20% C and for 5% Cr steels with and without 0.50% Mo. This strengthening effect of Mo at high temperatures will be discussed more fully in Chap. 19. There may be a similar strengthening effect in austenitic steels, but this is not so thoroughly proved.

**Corrosion and Oxidation Resistance.** Molybdenum added to austenitic Ni-Cr steel produces marked resistance to certain types of corrosion. In the paper industry 18-20 Cr, 8-10 Ni plus 2 to 4% Mo is standard for resisting corrosion by sulphite liquor. (See Chap. 20.) In pearlitic steels, however, Mo does not confer corrosion resistance or oxidation resistance. Indeed, steels high in Mo not only oxidize readily, but are peculiarly prone to decarburization, as will be discussed in Chap. 23. Chromium or Si is used to give the oxidation resistance needed for most high-temperature service, and the straight C-Mo steel without grain-size control additions is reserved for uses in which strength at high temperature is the primary requirement. In such cases about 0.50% Mo is the normal content, though it sometimes rises to 1.5%. When W is used to confer high-temperature strength, not less than 1% is required to give as much effect as 0.50% Mo.

In petroleum cracking tubes for use on non-corrosive crude oils, and under conditions where scaling is not rapid, the plain C-Mo steels find use, as they do in boiler tubes and the like, for super-heated steam service under high temperature and pressure. Although the enhance-

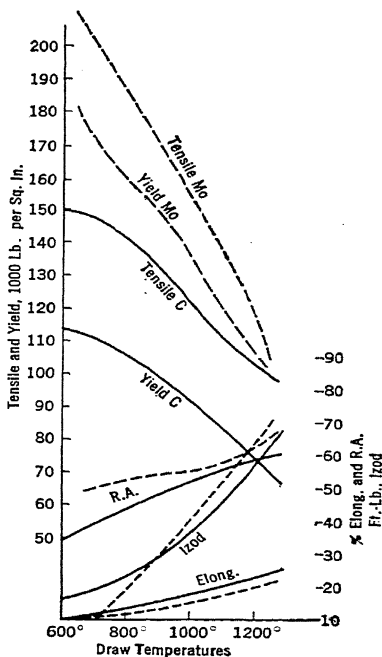


FIG. 5.—Properties of S.A.E. 1045 water quenched from 1500° (solid lines), and 0.39 C, 0.19 Si, 0.71 Mn, 0.36% Mo steel (dashed lines), water quenched from 1550°, drawn as shown.

(Data for 1045 from Nickel Alloy Steels,<sup>9</sup> for Mo steel from Molybdenum in 1933.<sup>10</sup>)

See general comment on property charts, p. 40, chap. I.

ment of high-temperature strength by Mo is very marked, strength at ordinary temperatures is not so notably affected. Such steel, of course, has more depth-hardening ability than plain C steel, but where this is sought it is ordinarily more economical to use a combination of more of some cheaper element with only say 0.20% Mo, than it is to raise the Mo to the 0.50–1.00% that might be necessary alone.

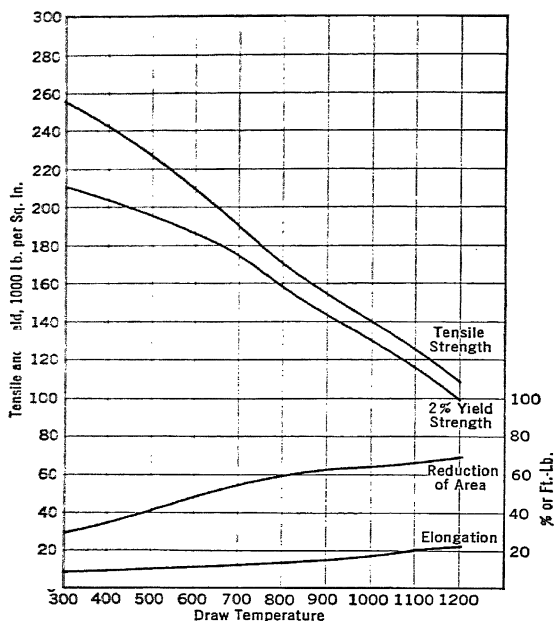


FIG. 6.—Average properties of "Amola" steel,<sup>8</sup> 0.25–0.35 C, 0.20–0.30 Si, 0.70–0.90 Mn, 0.15–0.25% Mo. Normalized from 1600°, water quenched from 1525°, drawn as shown. Treated in 0.53-in. diam., tested in 0.505-in. diam.

After curves copyrighted by Chrysler Corp.

See general comment on property charts, p. 40, chap. 1.

**Quenched and Tempered Constructional Molybdenum Steels.** The physical property chart for a water-quenched, fine-grained steel of 0.39 C, 0.36% Mo compared with that for water-quenched 1045, Fig. 5 shows that the alloy increases tensile and yield strengths and increases the reduction of area, but drops the elongation slightly. The impact value, at draw temperatures over 800°, is superior to that of the C

steel. Data for water-quenched 0.30 C, 0.25 Si, 0.80 Mn, 0.20% Mo steel, grain size controlled, are shown in Fig. 6.\*

The plain C steels with about 0.20% Mo which have been given small grain size by use of Al are used for springs as well as for general

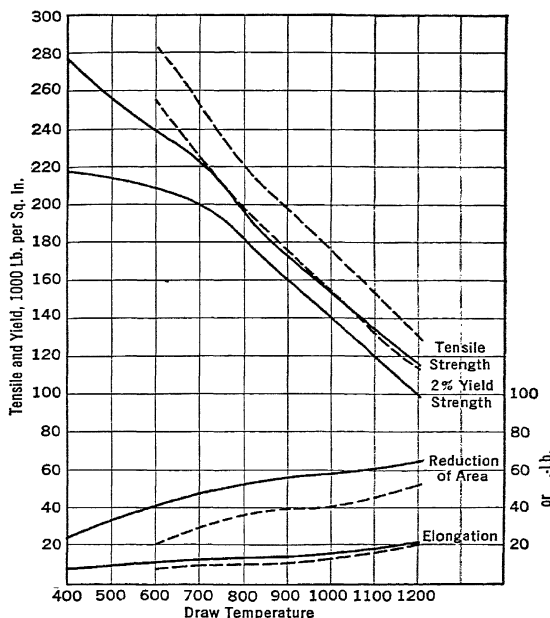


Fig. 7.—Average properties of “Amola” steel, solid lines for 0.35–0.45 C, 0.20–0.30 Si, 0.70–0.90 Mn, 0.15–0.25% Mo, normalized from 1600, oil quenched from 1500°, drawn as shown. Dashed lines for 0.60–0.70% C, other elements same as above, normalized from 1540°, oil quenched from 1450°, drawn as shown. Treated in 0.53-in. diam., tested in 0.505-in. diam.

After curves copyrighted by Chrysler Corp.  
See general comment on property charts, p. 40, chap. 1.

constructional steels. The usual range of C is 0.40 to 0.65%. Properties for these C contents in a 0.20% Mo steel, obtained on speci-

\*The data here given and those of Figs. 6 and 7 are taken, by permission of C. H. Wills, from curves copyrighted by the Chrysler Corporation and show averages of properties of “Amola” steel from a cooperative study made by the Chrysler Corporation and six alloy steel producers.

mens quenched in 0.53 in. diam. (this small section being chosen because of the applicability to springs which are usually of small section) are shown in Fig. 7. Data (not plotted) for tensile and yield of a 0.50 C, 0.20% Mo steel, normalized from 1600°, then oil quenched from 1475° fall between the curves for 0.40 and 0.65% C, the elongation falls close to that for 0.40% C, and the reduction of area is much closer to the curve shown for 0.40% than for 0.65% C. These curves may be compared with data for other spring steels given in Chap. 16.

The following extremes and averages are given by two classes of steels heat treated in 0.53 in. rounds normalized from 1600° (30 min.), heated to 1700° in cast-iron chips 1 hr., oil quenched, drawn 1 hr, at

C %	Si %	Mn %	Mo %	Tensile	Yield	Elong. %	R.A. %
0.15-0.25	0.20-0.30	0.70-0.90	0.15-0.25	180,000	135,000	10	25
				170,000	150,000	12.7	34.4—av.
				150,000	105,000	25	45
0.25-0.35	0.20-0.30	0.70-0.90	0.15-0.25	220,000	150,000	9	24
				190,350	134,250	11.1	30.9—av.
				170,000	120,000	37	13

Charpy impact values for a 0.65 C, 0.20% Mo spring steel on specimens heat treated in 0.394 in. square are shown in Fig. 8. Because

Charpy Impact, Ft. 40  
30  
20  
10

600° 800° 1000° 1200°  
Tempering Temperature

FIG. 8.—Charpy impact of steel of 0.65 C, 0.25 Si, 0.80 Mn, 0.20% Mo, oil quenched in 0.394 in. square bar, from 1450°, drawn as shown.

of the grain control exercised by the use of Al, these steels are not sensitive to quenching temperature. Twenty-seven specimens of a 0.65 C, 0.20% Mo steel were oil quenched in 0.46 in. diam., nine each from 1450°, 1500°, and 1550°, all drawn at 800°. The averages were as follows, all specimens in any group showing close agreement.

Quench Temp.	Rockwell C as Quenched	Tensile	Yield	Elong. %	R.A. %	Brinell after Draw
1450°	63	234,500	220,000	11	39.5	434
1500°	63	233,500	215,500	11	41	437
1550°	63	231,000	213,000	11	41	439

**Depth Hardening.** The bulk of the evidence as to depth hardening of Mo steels is best presented in relation to the complex steels (Chap. 13). It might here be noted that the hardenability of a 0.30 C, 0.50% Mo cast steel is so increased by the Mo that in welding cast valves of that composition onto low C-Mo steel tubes it is essential, according to the Welding Handbook,<sup>2</sup> to preheat before welding and to heat after welding for stress relief anneal.

**High Molybdenum Steel.** The properties of a 0.10 C, 0.12 Si, 0.28 Mn, 1.38% Mo steel are given<sup>3</sup> as follows:

	TENSILE	YIELD	ELONG. %	R.A. %	CHARPY
As Received.....	75,000	43,000	33	69	108
Ann. 1650°,					
Drawn 1300°..	64,500	43,000	36	73	145

The steel was not air hardening. Creep data are given on this steel, and one with 1% Mo, for specimens whose surface had been "calorized" (aluminum coated) to combat oxidation. Increasing Mo from 1 to 1.5% increased the creep resistance.

**Molybdenum in Complex Steels.** There are no plain Mo or W Steels in the S.A.E. series. There is a considerable tonnage of S.A.E. Ni-Mo and Cr-Mo bar stock as well as some of Ni-Cr-Mo, the amount of Ni-Mo steels produced being as great as that of plain Ni, and the total tonnage of S.A.E. steels alloyed with Mo being greater than that of those alloyed with V, according to the estimates of one producer. As is the case with V, Mo finds large use as an addition to Cr steels. The Cr-Mo and Cr-V steels are discussed in Chap. 13. In cast steels, also, Mo is very rarely used alone. Its application is almost invariably in conjunction with other elements, though some C-Mo castings, without other alloy, are used in high-temperature turbine and other steam service, as well as in the oil-cracking industry.

The Mo and W steels are discussed in detail in two books by Gregg: "The Alloys of Iron and Molybdenum"<sup>4</sup> and "The Alloys of Iron and Tungsten"<sup>5</sup> and the Climax Molybdenum Co. has recently published a convenient summary of the properties of some Mo steels.<sup>11</sup>



## PLAIN MOLYBDENUM AND TUNGSTEN STEELS

### BIBLIOGRAPHY

1. Allowable Working Stresses for Ferrous Materials, *Mech. Engng.*, v. 60, 1938, pp. 170-171.
2. American Welding Society: "Welding Handbook." New York, The Society, 1938. 1210 pages.
3. CLARK, C. L., and R. S. BROWN: A New Low-Alloy Pearlitic Steel for High-Temperature Service, *Trans. Am. Soc. Mech. Eng.*, v. 59, 1937, pp. 541-544.
4. GREGG, J. L.: "The Alloys of Iron and Molybdenum." New York, McGraw-Hill Book Co., 1932. 507 pages.
5. GREGG, J. L.: "The Alloys of Iron and Tungsten." New York, McGraw-Hill Book Co., 1934. 511 pages.
6. EPSTEIN, S., J. H. NEAD, and J. W. HALLEY: Choosing a Composition for Low-Alloy High-Strength Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 120, 1936, pp. 309-345. Also: *Metals Technology*, v. 3, 1936, Tech. Pub. 697, 31 pages.
7. HADFIELD, R. A.: Addendum on Alloys of Iron and Molybdenum, *Proc. Am. Inst. Mech. Eng.* (1915), pp. 701-713.
8. GUILLET, L.: Comparaison des propriétés et classification des aciers ternaires (Comparison of the Properties and Classification of the Ternary Steels), *Rev. de Mét. Métall.*, v. 2, 1905, pp. 350-367. Tr.: *Stahl u. Eisen*, v. 25, 1905, pp. 1439-1444.
9. FRENCH, H. H., and J. W. SANDS (editors): "Nickel Alloy Steels." New York, International Nickel Co., Inc., 1934. 7 Sections.
10. Climax Molybdenum Company: "Mo-lyb-den-um in 1933." New York, The Author, 1933. 50 pages.
11. Climax Molybdenum Company: "Molybdenum in Steels." New York, Author, 127 pages.

## CHAPTER 10

### COMPARISON OF ALLOYING EFFECTS: S.A.E. STEELS AND TREATMENTS

**General Considerations.** In previous chapters the outstanding effects of the chief alloying elements have been briefly described. The range of properties conferred upon C steels by any single element is limited, and it is necessary to blend two or more alloying elements into more complex steels if we are to pass beyond these limits. Moreover, the alloying elements vary in cost and economic factors urge replacement of the more expensive ones by the cheaper where sufficiently identical properties may thus be had. A general recapitulation of the effects of the common alloying elements and the factors involved in using them together may be made before passing on the consideration of specific combinations of elements.

**Composition Hardness.** A rough-and-ready empirical comparison of alloy steels *in the same condition of heat treatment* is given in "Bethlehem Alloy and Special Steels"<sup>1</sup> for the calculation of "composition hardness." If the percentage of each alloying element is multiplied by an empirical factor and the products added, the stronger steel will usually show the greater sum. The method is applicable only for comparison of steels that are of the same general type, but the factors afford a sort of summary evaluation of the strengthening effect of the alloying elements.

Each 0.01% of a given element has, according to this source, the following factor, an index of its strengthening effect, by which its percentage is multiplied in the calculation.

P .....	40	Cr .....	5
C .....	30	Si .....	5
V .....	20	W .....	4
Mo .....	16	Ni .....	4
Mn .....	8	Cu .....	4

The Ryerson Company<sup>2</sup> calculates hardenability on quenching in a similar way, with practically the same factors. Burns, Moore, and

Archer<sup>2</sup> appraise hardenability on quenching in similar fashion, using the factors shown below:

C.....	1000
Cr.....	500
Mn.....	400
Ni.....	100
Cu.....	100

They do not count Si, S, or P in the usual amounts. However, if S is over 0.03% it reduces the amount of Mn available for hardening.

**Amounts Commonly Used.** The percentages of the different elements ordinarily utilized bear some relation to the "composition hardness," and may be summarized as follows:

- P — under 0.04% in heat-treated steels, around 0.10% in some free-machining steels, up to 0.15% in low-C, high-yield-strength steels.
- S — under 0.055% in heat-treated steels, 0.10–0.30% in some free-machining steels in conjunction with high Mn. (Not an alloying element.)
- C — 0.10–0.20% in carburizing, and 0.07–0.15% in high-yield-strength steels for welding; 0.20–0.40% in most constructional steels and castings; 0.40–1.20% in various spring and tool steels.
- Mn — Over 0.25% and under 2% in most structural steels. 12–14% in austenitic castings.
- Si — Very low in rimming steels, from 0.10–0.30% in most S.A.E. steels, around 0.50% in castings, up to 0.75% in high-yield-strength steels, 1–1.75% in graphitic steels, up to 2% in spring steels, still higher in low-C steels for electric uses.
- Cu — About 0.20% in "copper-bearing" atmospheric-corrosion resistant steels, up to 1% in high-yield-strength steels, 1 to 2% in steels for precipitation hardening.
- Ni — Usually 2.50–3.50% when used alone, and up to 5% in carburizing steels, 0.50–3.75% in complex steels. 8% or more added to corrosion-resistant Cr austenitic steels. Special uses for higher Ni steels.
- Cr — Under 1.50% in most structural steels for heat treatment. 0.50–4% in tool steels, larger amounts in still-tube steels, 12–30% in corrosion-resistant "stainless" steels.
- Al — 0.01–0.10% for grain-size control. Seldom used as ordinary alloying element, save 1–5% in special nitriding and up to 5% in special heat resisting steels.

- Ti — 0.01–0.20% for grain-size control.  
V — 0.01–0.20% for grain-size control and slight alloying effect.  
0.25–5% in tool steels.  
Mo — About 0.20% in constructional steels, smaller amounts in high-yield-strength steels; 0.50–1.50% in steels for high temperature service, up to 9% in tool steels.  
W — Very rarely used in constructional steels. Around 1–2% in steels for high-temperature service. 0.50–20% in tool steels.

Except for the auxiliary elements Al, V (and for Mo and W in steels for high-temperature service) the amount of any one element will be reduced when others are added. This is especially true for C. This is partly due to the effect upon the eutectoid ratio.

**Ferrite Formers and Carbide Formers.** In substitution, ferrite formers may replace other ferrite formers, one carbide former is used to replace or to supplement and intensify another, and ferrite formers and carbide formers are used to augment each other. In general, the ferrite formers give toughness, the carbide formers hardness, as will be brought out later in this chapter.

Classifying them, we have as pure ferrite formers with no entrance into carbides P—Si—Cu—Ni—Al. All the carbide formers (even C itself to a small degree, see Fig. 7, Chap. 2, Vol. I) are also soluble in ferrite, and hence divide between the ferrite and the carbide, the partition depending on the amount of C present, and, in the presence of C, to some extent on the temperature to which austenite has been heated in order to dissolve free carbides.

Manganese is the weakest carbide former and acts largely by strengthening ferrite. Molybdenum, Cr, W, and V are stronger carbide formers but can also enter ferrite. Of these the Cr and V carbides are the hardest and most wear resistant, the Mo and W carbides the most stable at elevated temperatures, i.e., they give more "red hardness" to high-speed tool steels.

**Influences on Austenite.** Solution of either complex carbides or non-carbide-forming elements in austenite modifies the behavior of the austenite. Some alloying elements limit the formation of austenite and produce the so-called "gamma loop" outside of which no inversions take place. Too large an addition of a loop-forming element will remove all possibility of heat treatment. The critical temperatures on heating are shifted either up or down, thus altering the heat-treatment temperature required. Addition of difficultly soluble carbides may call for high temperature and longer time for saturation and homogenization. The alloying elements tend to alter the grain growth and

coarsening temperature characteristics. The alloyed austenite is more sluggish, transforms less readily on cooling, and so has a slower critical cooling rate for complete hardening and a greater tendency toward depth hardening. That is, for C contents that require water hardening in C steel, a little alloy will usually make the steel oil hardening, more will make it air-hardening, and a lot may so lower the transformation temperature as to make the steel austenitic at room temperature even at the slowest rates of cooling.

**Depth Hardening.** On solution in austenite even the ferrite formers tend to make austenite sluggish, to lower the critical cooling rate, and confer depth-hardening properties; but in this respect the carbide formers, Mn, Cr, Mo, and W, are the most potent while V, in the small amounts used, has a minor but still an important effect. In order to get full effect from the depth-hardening power of the carbide formers,  $Ac_3$  must be materially exceeded in heating for quenching. Depth-hardening effects will be discussed a bit later. Heat-treatment temperatures are directly influenced by the change in  $Ac_3$  produced by alloying.

**Other Effects.** The presence of an alloying element, especially one that forms a stable carbide, may render martensite reluctant to temper and require higher drawing temperatures to soften it.

The alloying elements may alter the percentage of C required to form the eutectoid.

The ferrite formers, Si and Ni, also tend mildly to restrain grain growth, though the low-C, high-Si transformer steels with compositions beyond the gamma loop readily grow very coarse grains. In large amounts, Al, too, no longer acts as an inhibitor, but promotes coarseness.

From the above we see that with one group of elements we can strengthen ferrite, with another we can harden and stabilize the carbide, and, with some, we can strongly induce depth hardening.

**Availability.** Although experience has led to the adoption, as S.A.E. steels, of a very limited number out of the permutations and combinations possible with these and other alloying elements, and few grades beyond these are commercially available without a special order and accompanying higher cost than if they were in regular production, other combinations may be needed. If they are sufficiently studied, new combinations may prove to be as effective and cheaper than the established grades. So, while giving the present complex commercial steels most of our attention in subsequent chapters, we must outline some of the possibilities in steels not now regularly on the market.

The steel founder, with his ability to make small special heats in the electric furnace and with specifications based on mechanical properties rather than chemical composition, is especially well prepared to utilize new alloy combinations. Moreover, since the properties of cast steels reflect those of wrought steels of the same composition, it is of little moment whether the data available on compositions that are new, or not yet widely used, be from tests on cast or wrought steels.

**Recapitulation of Effects.** Listing the effects, we may consider:

1. Ferrite-strengthening and carbide-forming effects, grain-growth tendencies.
2. Critical-temperature and gamma-loop effects.
3. Tendency toward making austenite sluggish on cooling-depth hardening.
4. Tendency toward making martensite sluggish on tempering.
5. Effect on eutectoid C content.
6. Specific effects conferred by alloying elements.

**Ferrite-strengthening and Carbide-forming Effects.** A solid solution of two metals is usually stronger than either pure metal. The elements that do not form carbides obviously can act only as ferrite strengtheners; but those that form carbides can and do dissolve in ferrite as well as enter the carbide, though little is known about what governs the exact partition between ferrite and carbide. Of the common elements, the non-carbide formers which presumably enter the ferrite only are Ni, Co, Cu, and P. Silicon and Al form carbides under suitable conditions, but these conditions probably do not obtain in steel, these elements probably entering the ferrite chiefly, when present in normal amounts. Those that form carbides readily and divide between the carbides and the ferrite are Mn, Cr, Mo, W, V, Ti, and Cb. Some of these can form several different carbides.

Silicon carbide is readily formed at high temperature, but Si is not ordinarily considered to enter into  $\text{Fe}_3\text{C}$ . Aluminum can be made to form carbides, but, as in the case of Si, it is doubtful if these are formed in steel. Nickel carbides are known, but are not thought to be formed in steel. Carbides of Cu or P are unknown.

Manganese readily forms the sulphide and the oxide, so the first additions of Mn combine with these non-metals, if they are present in active form. Any excess tends to form  $\text{Mn}_3\text{C}$ , and doubtless the  $\text{Fe}_3\text{C}$  of all commercial steels is really some mixture of the two carbides in unknown and probably variable proportions,  $(\text{Fe}, \text{Mn})_3\text{C}$ . The special carbides present in some Cr steels have been mentioned in Chap. 7.

2 of the other carbide formers, when present in large amounts, as in high-speed steel, form a different complex carbide with only half as much C. (Fe, W, Mo, Cr, V)<sub>3</sub>C, but, when present only in small amounts, the carbides of all these metals are thought to be present in solution in Fe<sub>3</sub>C.

When the C is low, the amount of carbide-forming elements taken up by Fe<sub>3</sub>C is probably not large, and in such steels, as in the carbon-iron binary alloys, these elements chiefly enter into and strengthen the ferrite. This effect for Mn was pointed out long ago in Campbell's <sup>4</sup> formulae for tensile strength and is the basis of the formulae for composition hardness. Specific effects of the carbide formers will appear when we study the complex special-purpose steels in later chapters.

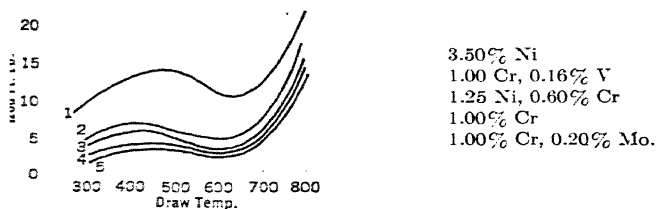


Fig. 1.—Behavior in impact of oil-quenched alloy steels of 0.50% C, drawn as shown. (Grossmann.<sup>5</sup>)

The toughening effect of the ferrite formers, as contrasted with the carbide formers can be shown by impact data given by Grossmann.<sup>5</sup> Fig. 1, for 0.50% C steels in the low range of draw temperatures. First compare steel No. 1 (with high Ni, a ferrite former) and No. 4 (with Cr, a carbide former). The former is the tougher. Number 3 with less Cr and a little Ni is more like the straight Cr; No. 5, with Cr and Mo, both carbide formers, is still less tough. Of course the toughness has to be considered in relation to the strength so that when everything is considered the stronger Cr-Mo is not inferior to the plain Cr.

Number 2, Cr-V is of interest in comparison with No. 4. Although V is a carbide former, its effect in producing fine grain size has overbalanced the de-toughening effect of Cr.

**Grain Size Control.** Cementite which carries the carbide-forming elements seems to dissolve less readily in austenite, and, since austenite is considered not to show marked grain growth till it has dissolved and diffused all the carbides, the result of the alloying is a finer grained steel in the as-rolled or normalized condition. The low percentage of V used in these and in most V-bearing steels (aside from high-speed

steels) indicates that it is employed more for grain-size control than for other alloying effects. Chromium has a slight effect in inhibiting grain growth, but Mn has practically no such inhibiting effect.

As has been mentioned in Chap. 8, Al, Ti, and Zr, even when used in amounts under 0.10%, all have marked tendencies toward inhibition of grain growth. Possibly the carbides of Ti and Zr are difficultly soluble like those of V\* and Mo, but since both readily form oxides and nitrides it is difficult to tell just what the mechanism is. With Al, the consensus of opinion is that the formation of  $\text{Al}_2\text{O}_3$  or some more complex Al-O compound, e.g., a silicate, is involved in the process by which Al tends to inhibit coarsening of the austenite grain. This does not exclude the possibility that other Al compounds beside the oxide are involved. Notwithstanding the small percentages used, these elements might be classed among the alloying elements.

**Effects Upon Yield Strength.** Data selected by Epstein, Nead, and Halley<sup>6</sup> as representative of the effect of C, Cu, Mn, Ni, P, Si, Cr, and Mo in Fe or very low-C steels have been given in the figures under those headings in previous chapters. These variations in strengthening effect, especially upon the yield strength, have been influential in the selection of alloying elements for a large series of so-called "low-alloy, high-yield-strength" steels designed to give in the as-rolled, or at least the merely normalized condition, a combination of high yield strength and ductility not attainable in C steel without quenching. This series will be discussed in the next chapter.

**Formation of Gamma Loops.** Some alloying elements, above a certain percentage (which is affected by the C content), entirely remove the possibility of heat treatment by quenching and tempering, or of grain refinement by normalizing, by making the steel ferritic at all temperatures below the freezing point. That is, the austenite or gamma field is constricted and forms the "gamma loop." The gamma fields for carbonless, or low-C, Fe-P, Fe-Mo, and Fe-Cr alloys are shown in Fig. 3, Chap. 2, Fig. 2, Chap. 9 and Fig. 3, Chap. 7. The sagging loop of the Fe-Cr system, in which the alpha to gamma critical temperature on heating first falls and then rises with increasing Cr, is rare, most gamma loops having the general contour shown in Fig. 3, Chap. 2, though the extent of the loop of course varies with the particular element. The presence of sufficient C takes the sag out of the Fe-Cr loop and extends it to higher Cr contents, as Fig. 3, Chap. 7 shows.

**Critical Temperatures.** Besides Mo and Cr, the elements W, V, P, Si, and Al all form gamma loops with Fe, whereas Ni, Cu, and Mn

\*Recent work by Lauderdale and Harder<sup>26</sup> challenges the generally accepted ideas of slow solubility of carbides. See also Bain.<sup>27</sup>



do not. The last three elements lower the critical temperature on heating. Cr, up to some 10% Cr, in C-free or low-C alloys, lowers it, larger amounts raise it, and Mo, W, V, P, Si, and Al all raise it. In complex steels the effect of one element in altering the critical temperature may be opposed or increased by the effect of another.

Within the ranges met in most S.A.E. steels

0.1% C lowers  $A_{c3}$  approximately  $4^\circ$ .

0.01% Mn lowers  $A_{c3}$  approximately  $0.5^\circ$ .

1% Ni or Cu lowers  $A_{c3}$  approximately  $0.4^\circ$ .

1% P raises  $A_{c3}$  approximately  $8^\circ$ .

1% Si raises  $A_{c3}$  approximately  $0.5^\circ$ .

0.01% V raises  $A_{c3}$  approximately  $0.6^\circ$ .

1% Mo raises  $A_{c3}$  approximately  $1.3^\circ$ .

0.01% W raises  $A_{c3}$  approximately  $0.5^\circ$ .

The effect of Cr varies with C content.

The lowering of  $A_{r1}$  on slow cooling is of course a function of the rate of cooling as well as of the content of C and of alloying elements.

The important critical temperatures for heat treatment are  $A_{c3}$ , the completion of the critical changes on heating, i.e., complete formation of austenite in hypoeutectoid steels, and  $A_{r1}$ , the eutectoid temperature for completion of the formation of pearlite from austenite on slow (i.e., furnace) cooling. The data given above show that steels containing V or Mo, or Si or P above the normal amounts, will require higher temperatures to become austenitic, that the effect of Cr varies according to what else is in the steel, and that the presence of Mn, Ni, or Cu will allow the use of lower temperatures, whereas a complex steel, containing elements of each of the two classes, will take a temperature that is the resultant of the separate effects.

Morris, Sergeson, and Gable<sup>7</sup> give the critical temperatures on heating and on furnace cooling for the S.A.E. steels. A few plots of selected data are given in Fig. 2 to aid in visualizing the effect. Of course the C content is the main variable. It will be seen that Mn and Ni have a very strong depressing action, especially on  $A_{r1}$ , most marked in higher C steels, and that Cr, in low and medium C raises  $A_{r1}$ , and when in company with Ni it opposes the lowering effect of Ni. An element or combination of elements that lowers  $A_{r1}$  is likely to increase the depth of hardening, i.e., to lower the critical rate of cooling.

**Critical Cooling Rates—or Depth Hardening.** The critical cooling rates (see Chap. 6, Vol. I) to produce martensite free from troostite are roughly expressed, in everyday parlance, by the terms "water

hardening," "oil hardening," or "air hardening," rather than in terms of degrees per second. The relative hardenability of various alloy steels and of steels of various grain sizes can be quantitatively shown by quenching specimens of equal dimensions, sectioning, and plotting the

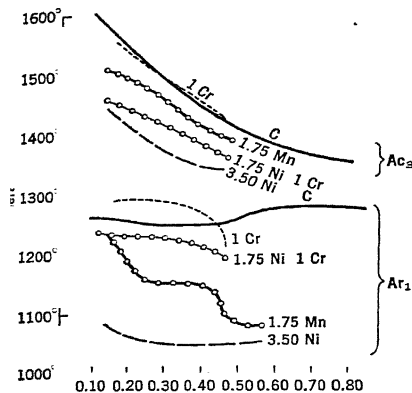
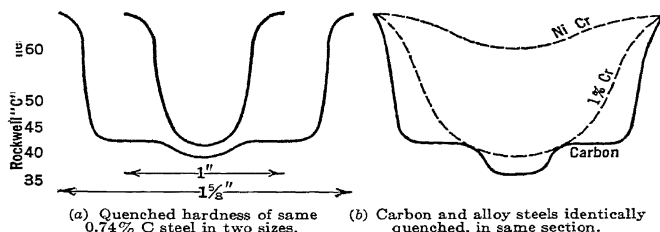


FIG. 2.—Effect of C and alloying elements upon critical temperature on heating and cooling. (After Morris, Sergeson and Gable.<sup>7</sup>)

drop of hardness from edge to center as has been mentioned in Chap. 9, Vol. I.

Carbon steel will harden only to around 0.25 in. depth no matter how rapid the quenching. This is nicely shown in Fig. 3a from Bain.<sup>8</sup>



(a) Quenched hardness of same 0.74% C steel in two sizes. (b) Carbon and alloy steels identically quenched, in same section.

FIG. 3.—Depth hardening in C and alloy steels. (Bain.<sup>8</sup>)

Many alloy steels will harden to a much greater depth even with oil quenching, and some highly alloyed steels will harden throughout in large masses even on air cooling. Figure 3b compares C steel with two alloy steels. This property of depth hardening is of great importance when large sections are to be heat treated, and maximum properties are desired at the core. The ability to make steel harden

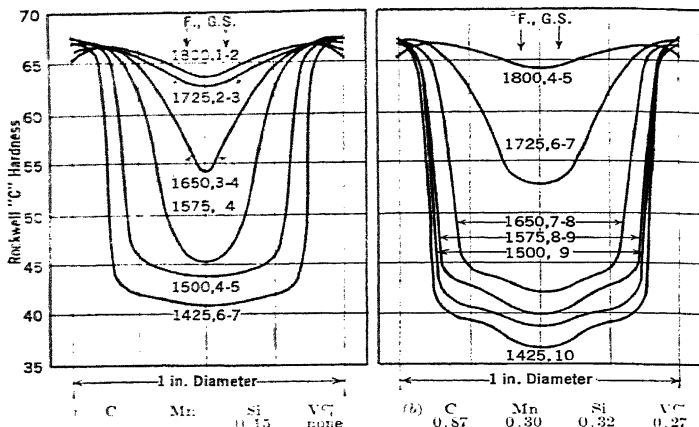


FIG. 4.—Hardenability as affected by successively higher quenching temperatures. Steel brine quenched, not drawn. Heating temperatures for quenching and A.S.T.M. grain size produced at each temperature are shown on the curves. (Zimmerman, Aborn and Bain.<sup>25</sup>)

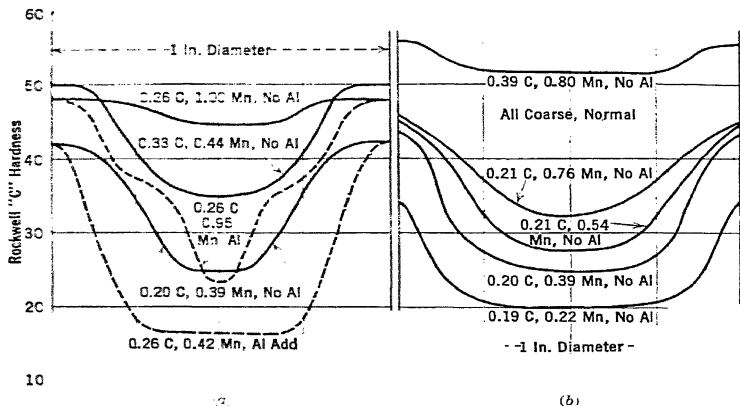


FIG. 5.—Effect of varying Mn and C, and of producing fine grain, upon hardenability. Steels water quenched from 1700°, not drawn. (After McQuaid.<sup>9</sup>)

at slow cooling rates, i.e., to harden deeply in large masses, is one of the most potent reasons for the use of alloys in steel. The different alloying elements vary widely in their ability to confer depth hardening.

**Grain Size.** First of all, let us recapitulate the effect upon depth hardening resulting from grain size. Figure 4 shows the differences in hardenability concomitant with grain growth in C and V steels quenched from different temperatures. McQuaid<sup>9</sup> brings out the ef-

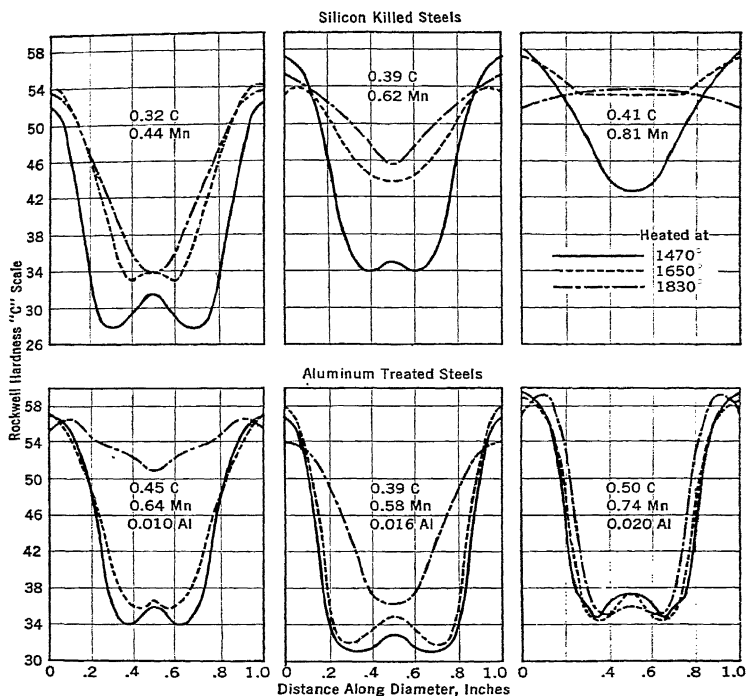


Fig. 6.—Hardness penetration in 1-in. round bars heated to various temperatures for 1 hr. and water quenched from 1470°. (McBride, Herty and Mehl.<sup>10</sup>)

fect of Mn and C in Fig. 5b and of coarse *vs.* fine grain in Fig. 5a. Figure 6 shows data from McBride, Herty, and Mehl.<sup>10</sup>

The presence of a soft center in even relatively small pieces of shallow hardening steels is clearly shown by these figures. The behavior of larger sections is shown by Knowlton,<sup>11</sup> Fig. 7 for steels of varying composition, quenched from proper hardening temperatures, as 3-in. diam. axles, and tempered back at suitable temperatures to

produce two levels of surface hardness, 475 and 340 Brinell, then sectioned and Rockwelled across a diameter.

Grossman<sup>7</sup> shows diagrams (Fig. 8) that again contrast the shallow hardening of C steel with the deeper hardening of suitable alloy steels on water quenching, and show how the two steels behave on oil

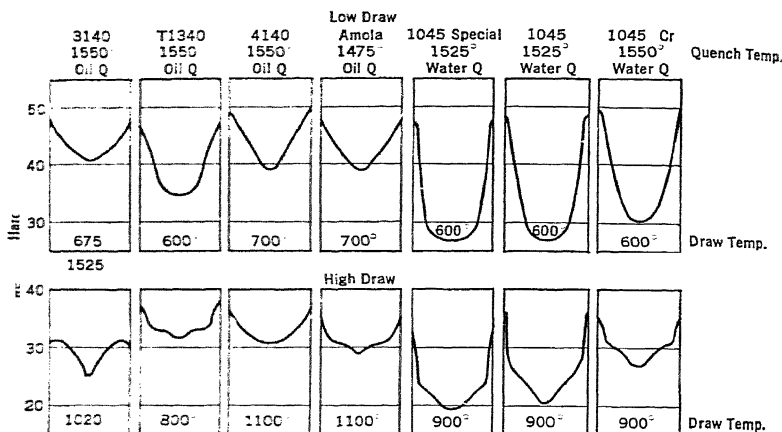


FIG. 7.—Penetration of hardness on 3-in. diam. stock.

No.	COMPOSITION			McQUAID EHN GRAIN SIZE	SURFACE BRINELL	
	Mn	M	Mo		Low Draw	High Draw
3140		1.17	0.74			
T1340			1.08	6-7	430	365
			0.27	7-8		
4140	0.71	0.23	0.10	5-7	475-495	330-340
1045	0.74		0.06	2-5		
1045 - Cr			0.27	2-5		

(Knowlton,<sup>11</sup>)

quenching, the C steel failing to become martensitic in oil even in small sizes.

**Molybdenum for Depth Hardening.** Jones<sup>12</sup> has shown the marked effect of Mo (Fig. 9). Greaves and Rees<sup>13</sup> recently studied the mass effects in the following steels:

	C	Si	Cr	Mn	Ni	Mo	
Ni-Cr.....	0.42	0.25	0.41	0.72	1.20	....	
Cr-V.....	0.43	0.28	1.20	0.60	0.13	....	0.24
Cr-Mo.....	0.34	0.18	1.20	0.57	0.12	0.26	
Mn-Mo.....	0.41	0.22	0.08	1.39	0.20	0.32	

These were oil quenched from 1545° and drawn at a range of tempera-

tures in various sizes. Bars of 1-in. to 3-in. diam. were 5 in. long, larger bars had a length 50% greater than the diameter. The hard-

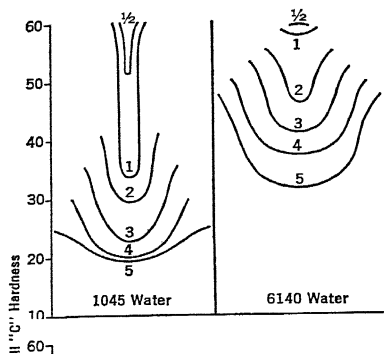
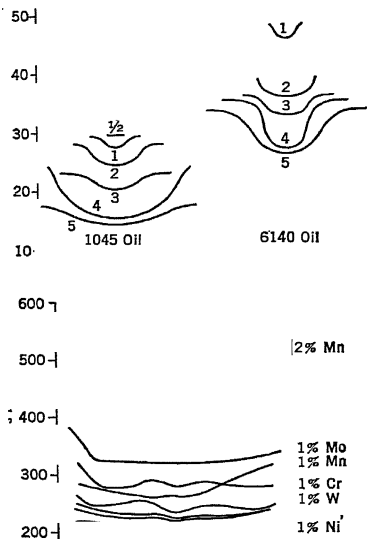


FIG. 8.—Size effect in quenched 1045 (0.45% C) and 6140 (0.40 C, 1.00 Cr, 0.16% V). Not drawn. Quenched in sizes ranging from  $\frac{1}{2}$ - to 5-in. diam. (Grossmann.<sup>5</sup>)



All steels contained  
0.33–0.36 C, 0.44–0.54 Mn,  
0.13–0.17% Si.  
Specimens were  $1\frac{3}{4} \times 1\frac{3}{4}$ -in. square  
bars, cast, not worked.  
All oil quenched from 1580°, not  
drawn.

-1.75 in.-

FIG. 9.—Depth hardening of alloy steels. (Jones.<sup>12</sup>)

ness for 1, 3, and 6-in. diam. bars is shown in Fig. 10. No curve is shown for the Ni-Cr steel since at 3 $\frac{3}{4}$ -in. diam. it had failed to show appreciable depth hardening.

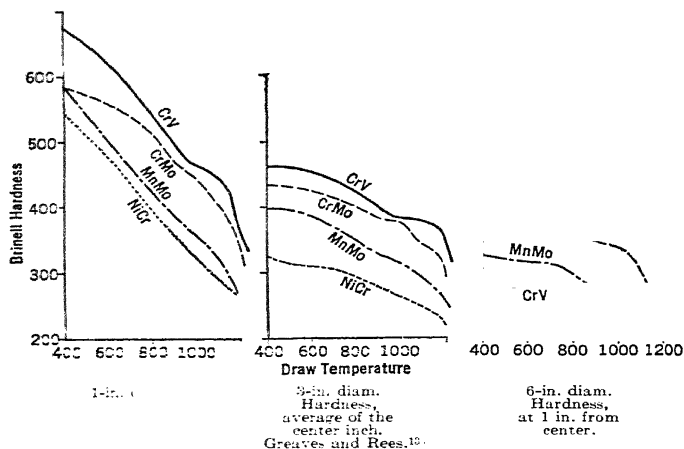


FIG. 10.—Mass effect in common alloy steels.

The properties of specimens from steels tempered 2 hr. at 1200° were:

		Tensile	Yield	Elong. %	R.A. %	Izod	Brinell
1-in. diam. bar.	Ni-Cr	124,000	101,000	23	62	70	263
	Cr-V	144,500	126,000	20	54	27	302
	Cr-Mo	134,500	114,500	21	63	64	289
	Mn-Mo	132,000	111,000	21	58	60	283
3-in. diam. bar	Ni-Cr	112,500	77,000	23	58	35	228
	Cr-V	129,000	105,000	20	56	39	269
	Cr-Mo	119,500	94,000	21	58	42	251
	Mn-Mo	121,000	91,000	21	55	33	251
6-in. diam. bar	Cr-V	center	103,000	42,000	27	61	207
		edge	126,000	98,500	21	57	264
	Cr-Mo	center	112,000	86,000	21	55	228
		edge	119,000	92,000	22	60	250
	Mn-Mo	center*	123,500	93,000	19	50	254
		edge	116,500	85,000	22	56	243

\* Carbon

ion suspected at center in Mn-Mo.

The depth hardening conferred by Mo is very marked.

Janitzky<sup>14</sup> gives Figs. 11 and 12 showing how Cr boosts the depth hardening in Ni steel. The potency of Mo is also indicated.

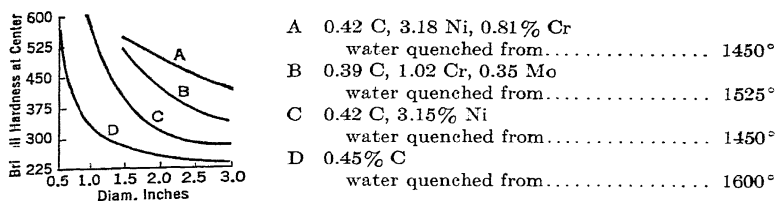


FIG. 11.—Hardness at center of bars ranging from  $\frac{1}{2}$ - to 3-in. diam. Water quenched, not drawn. (Janitzky.<sup>14</sup>)

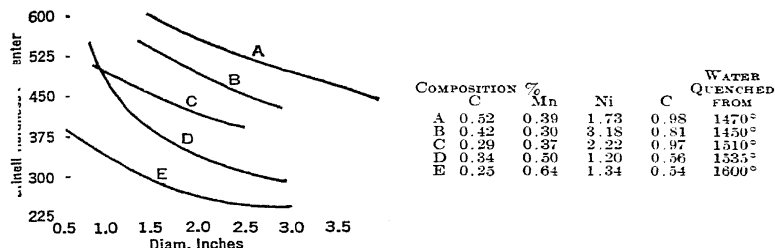


FIG. 12.—Hardness at center of bars of varying diameter, of NiCr steels varying in C, Ni, and Cr. Water quenched, not drawn. (Janitzky.<sup>14</sup>)

**Critical Cooling Rates.** Esser, Eilender, and Majért<sup>15</sup> record the following critical cooling rates for complete hardening in 0.40% C steel.

0.55% Mn.....	1000° per sec.
1.6 % Mn.....	90
2.2% Mn.....	15
1% Ni.....	800
3% Ni.....	325
5% Ni.....	150
1% Si.....	1400
1.5% Si.....	1800
0.50% Cr.....	725
1% Cr.....	180
2% Cr.....	40



Kallen and Schrader<sup>10</sup> have given data from which Fig. 13 can be drawn up for the surface hardness of specimens ranging from 0.75 in. to 8 in. diameter, quenched into warm water or warm oil (105°–125°). The upper line of each plot is for the 0.75-in. specimen, those below it

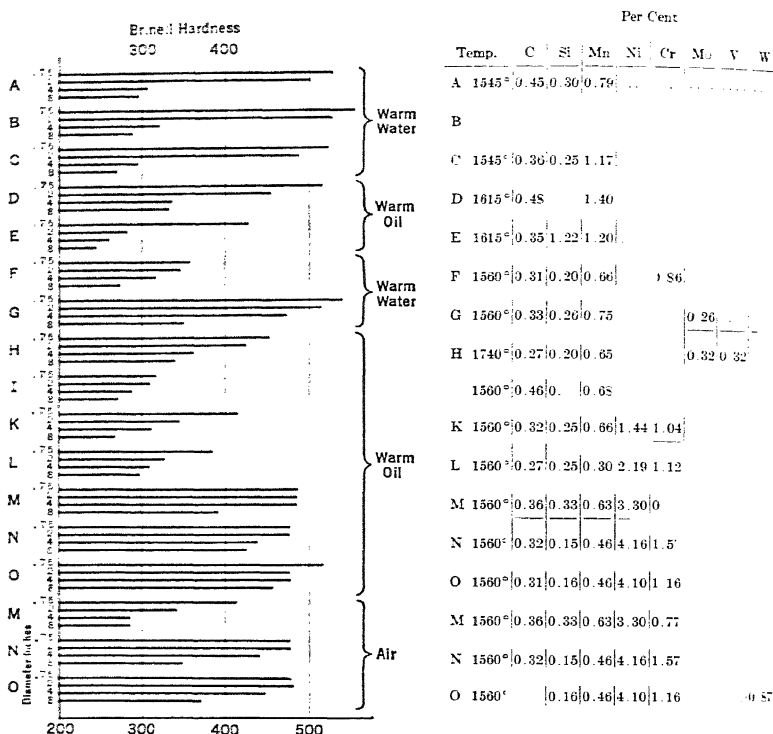


FIG. 13.—Mass effect on alloy steels.

Brinell hardness taken on outer zone of quenched specimen. Specimens of 0.75-, 2-, 4-, and 8-in. diam. quenched in warm 105°–125° water or warm oil, or in air, not drawn.  
After Kallen and Schrader.<sup>10</sup>

respectively for the specimens of 2 in., 4 in., and 8 in. diameter. Although the C content varies too much for a very close comparison and the hot water and oil are responsible for the 0.75-in. specimen failing to show real hardening in the case of some steels that would normally be classed as satisfactorily water or oil hardening, interesting com-

parisons are brought out. For example, comparing steels F and G, the potency of Mo in giving depth hardening is shown. That steel I did not harden is accounted for by the cooling rate figure of  $800^{\circ}$  per sec. cited above from Esser and co-workers. Steels M and N, quenched in oil, looked as though M, with its slightly higher C, does not require quite so rapid a cooling rate as N, but the data for air-hardened specimens, at the bottom of Fig. 12, show that the more highly alloyed steel does air harden the more strongly, especially in the very large sections.

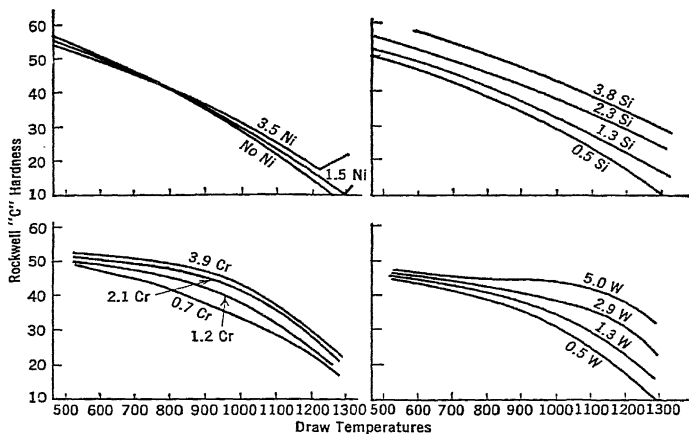


FIG. 14.—Effect of Ni, Si, Cr and W on resistance to tempering. (Bain.<sup>8</sup>)  
All steels contained 0.55% C. After quenching they were drawn for 1 hr. at temperatures shown.

In connection with depth hardening we should remember that liability toward cracking and toward production of internal stress increases with the severity of quenching and with the tendency to form martensitic shells on troostitic or pearlitic cores, so that ability to depth harden with a mild quench reduces these troubles. Hence oil- or air-hardening steels may be chosen even in the smaller sections when these difficulties arise. The effect of mass will be further discussed in Chaps. 12 and 13. The American Society for Metals held a symposium on hardenability of alloy steels in the fall of 1938. These papers are a useful source of collected information on the subject.<sup>28-33</sup>

**Effect of Alloying Elements on Softening of Martensite in Tempering.** The higher the temperature and the longer the time at which a hardened steel is tempered, other things being equal, the less the internal stress. This is reflected in improved fatigue properties.<sup>17, 18</sup>

alloying elements have in marked degree this property of stabilizing martensite and secondary troostite against softening. Outstanding for this purpose are Mo and W, this being in line with their effect on high-temperature strength. Silicon shows the effect to some degree; Cr has it markedly, but Ni only very slightly. Figure 14 collected from figures shown by Bain,<sup>5</sup> illustrates this. Bain states that Mo is fully as effective as W and ascribes the phenomenon to the formation of stable carbides. The effect of Si might be ascribed to its effect in slowing down diffusion of C, but Bain notes that it adds to the central (core) hardness and so concludes that the apparent effect is primarily due to ferrite strengthening. The stabilizing effect of Mo is shown in Fig. 15.

**Effect of Alloying Elements on Eutectoid Composition.** Reed<sup>12</sup> has studied the effect of added elements upon the eutectoid C composition. Confining the discussion to the amounts of alloying elements present in commercial steels for heat treatment, it can be shown from his data that, taking the straight Fe-C eutectoid at 0.89% C (which is probably 0.05 and perhaps 0.10% too high), the addition of the amount of alloying element stated below lowers the eutectoid C to the value shown:

V	0.2%	to	0.85%
V	1.0		0.65
Mn	1.0		0.77 <sup>c</sup>
	2.0		0.63 <sup>c</sup>
Cr	0.5		0.83
	1.0		0.78
	2.0		0.65
Mo	0.25		0.87
	0.5		0.85
	1.0		0.81
W	1.0		0.84
Ni <sup>a</sup>	1.0		0.85
	2.0		0.80
	3.0		0.76
	5.0		0.67
Si	1.0		0.83 <sup>b</sup>
	2.0		
	1.0		0.92 <sup>c</sup>

a. See also Fig. 7, Chap. 6.

b. Reed was unable to decide upon the effect of Si. These data are from Kriz<sup>13</sup> and Pobořil<sup>14</sup> and Scheil<sup>15</sup>.

c. These data from Showalter, Delamater, and Schwartz<sup>16</sup>.

According to data assembled by Houdremont<sup>17</sup>, Si reduces the eutectoid ratio very slightly more than Mn does, either Mn or Si having about double the effect of Ni.

Copper and Al, according to Reed, do not decrease the eutectoid C content and may slightly increase it. According to the data cited by Gregg,<sup>20</sup> the ternary eutectoid of Fe, Cu, and C is at 0.90% and 1.9% Cu.

If the effects are cumulative, a 1.25 Ni, 0.60% Cr steel would have a eutectoid C content of about 0.77% C or less, and a steel of 1.50 Ni, 0.20% Mo one of about 0.80% C or less, so that these carburizing steels will require approximately 0.10% less C to produce a hyper-eutectoid case than do the plain C steels, if the 0.89% C figure is accepted for the latter. Since Reed's basic assumption that 0.89% C

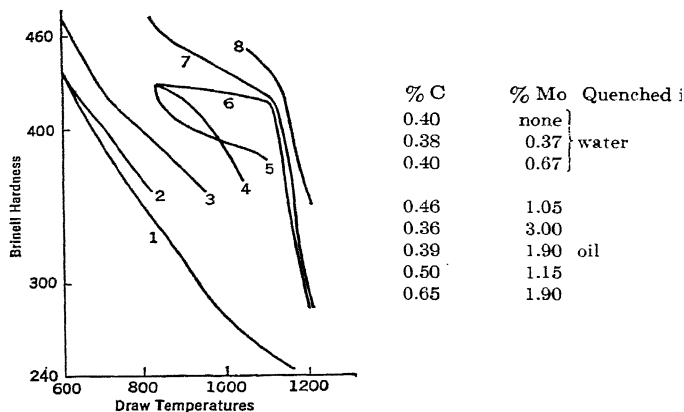


FIG. 15.—Resistance to tempering of steels of varying C and Mo content. (Gillett and Mack.<sup>17</sup>)

is the true figure for C steels is probably quite a bit too high, his figures cannot be accepted as quantitative, but his observations should give a rough indication of the effect. Thus it can be stated that the elements that are outstanding in producing depth hardening, Mn, Cr, Mo, and Ni, also shift the eutectoid C to a slightly lower figure.

Although, within limits, most of the elements are decidedly interchangeable, specific and individual effect do stand out for almost every alloying element, aside from its ability to substitute for others in general strengthening effect.

For example, one may turn to Mo for prevention of temper brittleness.

to Cr for wear-resistant carbides.

to Cr also (especially in large amounts) for corrosion resistance.

- to Ni for toughness and low-temperature impact resistance.
- to Cu for precipitation-hardening effect.
- to Cu and P for resistance to atmospheric corrosion.
- to Mo and to W for high-temperature strength.
- to W, Mo, and Co for "hot hardness."
- to V and to very small amounts of Al, Ti, or Zr, for grain-size control, and toughness due to small grain size.
- to Mn and Cr for cheap depth hardening.
- to Si for graphitization effects.
- to Ti and Cb for fixation of C.

Thus, when we want merely to meet a given set of tensile test specifications, we may balance toughness and hardness in various ways and produce a large range of practically equivalent steels; but, when we wish also to impart some specific effect, we may often find an alloying element that has special ability to produce it. On such considerations is based the procedure in making a modern "tailor-made" steel.

We may sum up the effects of the usual alloying elements as

< FERRITE	FORM CARBIDES
P, very strong	Ti, Cb, V, very strong
Si, Mn, strong	Cr, Mo, W, strong
Ni, Cu, Cr, Mo, W, fairly strong	Mn, fairly strong
V, weak	P, Ni, Cu, Si, no carbides
CONFER DEPTH HARDENING	STABILIZE MARTENSITE AND TROOSTITE ON TEMPERING
Mo, W, Mn, Cr, strong	Mo, W, strong
Si, medium	V, fairly strong
Cu, Ni, weak	Cr, weak
V, very weak	Mn, very weak
RESTRAIN GRAIN GROWTH OF AUSTENITE	Others, no appreciable stabilization
Al, very strong	DECREASE EUTECTOID CONTENT
V, Ti, Zr, fairly strong	Mn, Cr, V, considerably
Mo, medium	Ni, Mo, W, somewhat
Cr, weak	Si, Cu, Al, no appreciable effect
Si, weak	
Mn, very weak	
Ni, practically no action	

## SPECIFIC EFFECTS

Mo minimizes temper brittleness.

Mo, W, confer high-temperature strength especially in company with other carbide formers.

Ni confers toughness without brittleness, and especially confers low-temperature impact toughness.

Cu over about 0.80% confers ability to precipitation harden.

Cu and Cu + P, even in small amounts, confer improved resistance to atmospheric corrosion.

Cr in large amounts confers resistance to corrosion under oxidizing conditions and to oxidation on heating.

By balancing these effects, useful combinations are possible, some of which have found sufficiently wide use to warrant their listing in the S.A.E. series given below.

## S.A.E. STEELS AND TREATMENTS

**S.A.E. Steels.** The S.A.E. compositions for C and alloy steels are given below. For changes, additions, and further information, see the latest "S.A.E. Handbook."

The basic numerals for the various types of S.A.E. steel are

TYPE OF STEEL	NUMERALS AND DIGITS
Carbon Steels.....	1xxx
Plain Carbon.....	10xx
Free Cutting (Screw Stock).....	11xx
Free Cutting, Manganese.....	X13xx
High Manganese.....	T13xx
Nickel Steels.....	2xxx
0.50 Per Cent Nickel.....	20xx
1.50 Per Cent Nickel.....	21xx
3.50 Per Cent Nickel.....	23xx
5.00 Per Cent Nickel.....	25xx
Nickel Chromium Steels.....	3xxx
1.25 Per Cent Nickel, 0.60 Per Cent Chromium.....	31xx
1.75 Per Cent Nickel, 1.00 Per Cent Chromium.....	32xx
3.50 Per Cent Nickel, 1.50 Per Cent Chromium.....	33xx
3.00 Per Cent Nickel, 0.80 Per Cent Chromium.....	34xx
Corrosion and Heat Resisting Steels.....	30xxx
Molybdenum Steels.....	4xxx
Chromium.....	41xx
Chromium Nickel.....	43xx
Nickel.....	46xx and 48xx

TYPE OF STEEL	NUMERALS AND DIGITS
Chromium Steels.....	.5xxx
Low Chromium.....	.51xx
Medium Chromium.....	.52xxx
Corrosion and Heat Resisting.....	.51xxx
Chromium Vanadium Steels.....	.6xxx
Tungsten Steels.....	.7xxx and 7xxxx
Silicon Manganese Steels.....	.9xxx

**Prefixes.** The prefix 'X' is used in several instances to denote variations in the range of Mn, S, or Cr.

The prefix 'T' is used with the Mn Steels (1300 Series) to avoid confusion with steels of somewhat different Mn range that have been identified by the same numerals but without the prefix.

CARBON STEELS

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus Max.	Sulphur Max.
1010	0.05-0.15	0.30-0.60	0.045	0.055
1015	0.10-0.20	0.30-0.60	0.045	0.055
X1015	0.10-0.20	0.70-1.00	0.045	0.055
1020	0.15-0.25	0.30-0.60	0.045	0.055
X1020	0.15-0.25	0.70-1.00	0.045	0.055
1025	0.20-0.30	0.30-0.60	0.045	0.055
X1025	0.20-0.30	0.70-1.00	0.045	0.055
1030	0.25-0.35	0.60-0.90	0.045	0.055
1035	0.30-0.40	0.60-0.90	0.045	0.055
1040	0.35-0.45	0.60-0.90	0.045	0.055
X1040	0.35-0.45	0.40-0.70	0.045	0.055
1045	0.40-0.50	0.60-0.90	0.045	0.055
X1045	0.40-0.50	0.40-0.70	0.045	0.055
1050	0.45-0.55	0.60-0.90	0.045	0.055
X1050	0.45-0.55	0.40-0.70	0.045	0.055
1055	0.50-0.60	0.60-0.90	0.040	0.055
X1055	0.50-0.60	0.90-1.20	0.040	0.055
1060	0.55-0.70	0.60-0.90	0.040	0.055
1065	0.60-0.75	0.60-0.90	0.040	0.055
X1065	0.60-0.75	0.90-1.20	0.040	0.055
1070	0.65-0.80	0.60-0.90	0.040	0.055
1075	0.70-0.85	0.60-0.90	0.040	0.055
1080	0.75-0.90	0.60-0.90	0.040	0.055
1085	0.80-0.95	0.60-0.90	0.040	0.055
1090	0.85-1.00	0.60-0.90	0.040	0.055
1095	0.90-1.05	0.25-0.50	0.040	0.055

## S.A.E. COMPOSITIONS

## FREE CUTTING STEELS

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus Range	Sulphur Range
1112	0.08-0.16	0.60-0.90	0.09-0.13	0.10-0.20
X1112	0.08-0.16	0.60-0.90	0.09-0.13	0.20-0.30
1115	0.10-0.20	0.70-1.00	0.045 max.	0.075-0.15
1120	0.15-0.25	0.60-0.90	0.045 max.	0.075-0.15
X1314	0.10-0.20	1.00-1.30	0.045 max.	0.075-0.15
X1315	0.10-0.20	1.30-1.60	0.045 max.	0.075-0.15
X1330	0.25-0.35	1.35-1.65	0.045 max.	0.075-0.15
X1335	0.30-0.40	1.35-1.65	0.045 max.	0.075-0.15
X1340	0.35-0.45	1.35-1.65	0.045 max.	0.075-0.15

## MANGANESE STEELS

See Note (a) for Silicon Content

No.	Carbon	Manganese	Phosphorus	Sulphur
T1330	0.25-0.35	1.60-1.90	0.040 max.	0.050 max.
T1335	0.30-0.40	1.60-1.90	0.040 max.	0.050 max.
T1340	0.35-0.45	1.60-1.90	0.040 max.	0.050 max.
T1345	0.40-0.50	1.60-1.90	0.040 max.	0.050 max.
T1350	0.45-0.55	1.60-1.90	0.040 max.	0.050 max.

## NICKEL STEELS

See Notes (a) and (b) for Silicon, Phosphorus, and Sulphur Content

No.	Carbon	Manganese	Nickel
2015	0.10-0.20	0.30-0.60	0.40-0.60
2115	0.10-0.20	0.30-0.60	1.25-1.75
2315	0.10-0.20	0.30-0.60	3.25-3.75
2320	0.15-0.25	0.30-0.60	3.25-3.75
2330	0.25-0.35	0.50-0.80	3.25-3.75
2335	0.30-0.40	0.50-0.80	3.25-3.75
2340	0.35-0.45	0.60-0.90	3.25-3.75
2345	0.40-0.50	0.60-0.90	3.25-3.75
2350	0.45-0.55	0.60-0.90	3.25-3.75
2515	0.10-0.20	0.30-0.60	4.75-5.25



## 200 COMPARISON OF ALLOYING EFFECTS: S.A.E. STEELS

NICKEL-CHROMIUM STEELS  
See No. and *b* for Silicon, Phosphorus, and Sulphur Content

No.	Carbon	Manganese	Nickel	Chromium
3115	0.10-0.20	0.30-0.60	1.00-1.50	0.45-0.75
3120	0.15-0.25	0.30-0.60	1.00-1.50	0.45-0.75
3125	0.20-0.30	0.50-0.80	1.00-1.50	0.45-0.75
3130	0.25-0.35	0.50-0.80	1.00-1.50	0.45-0.75
3135	0.30-0.40	0.50-0.80	1.00-1.50	0.45-0.75
3140	0.35-0.45	0.60-0.90	1.00-1.50	0.45-0.75
N3140	0.35-0.45	0.60-0.90	1.00-1.50	0.60-0.90
3145	0.40-0.50	0.60-0.90	1.00-1.50	0.45-0.75
3150	0.45-0.55	0.60-0.90	1.00-1.50	0.45-0.75
3215	0.10-0.20	0.30-0.60	1.50-2.00	0.90-1.25
3220	0.15-0.25	0.30-0.60	1.50-2.00	0.90-1.25
3230	0.25-0.35	0.30-0.60	1.50-2.00	0.90-1.25
3240	0.35-0.45	0.30-0.60	1.50-2.00	0.90-1.25
3245	0.40-0.50	0.30-0.60	1.50-2.00	0.90-1.25
3250	0.45-0.55	0.30-0.60	1.50-2.00	0.90-1.25
3312	max. 0.17	0.30-0.60	3.25-3.75	1.25-1.75
3325	0.20-0.30	0.30-0.60	3.25-3.75	1.25-1.75
3335	0.30-0.40	0.30-0.60	3.25-3.75	1.25-1.75
3340	0.35-0.45	0.30-0.60	3.25-3.75	1.25-1.75
3415	0.10-0.20	0.30-0.60	2.75-3.25	0.60-0.95
3435	0.30-0.40	0.30-0.60	2.75-3.25	0.60-0.95
3450	0.45-0.55	0.30-0.60	2.75-3.25	0.60-0.95

MOLYBDENUM STEELS  
*a* and *b* for Silicon, Phosphorus, and Sulphur Content

S.A.E. No.	Carbon	Manganese	Chromium	Nickel	Molybdenum
4130	0.25-0.35	0.50-0.80	0.50-0.80		0.15-0.25
N4130	0.25-0.35	0.40-0.60	0.80-1.10		0.15-0.25
4135	0.30-0.40	0.60-0.90	0.80-1.10		0.15-0.25
4140	0.35-0.45	0.60-0.90	0.80-1.10		0.15-0.25
4150	0.45-0.55	0.60-0.90	0.80-1.10		0.15-0.25
4340	0.35-0.45	0.50-0.80	0.50-0.80	1.50-2.00	0.30-0.40
4345	0.40-0.50	0.50-0.80	0.60-0.90	1.50-2.00	0.15-0.25
4615	0.10-0.20	0.40-0.70		1.65-2.00	0.20-0.30
4620	0.15-0.25	0.40-0.70		1.65-2.00	0.20-0.30
4640	0.35-0.45	0.50-0.80		1.65-2.00	0.20-0.30
4815	0.10-0.20	0.40-0.60		3.25-3.75	0.20-0.30
4820	0.15-0.25	0.40-0.60		3.25-3.75	0.20-0.30

*a* included in the S.A.E. list, the trade refers to a steel of about 0.20 C, 0.65 Mn, 0.25% Mo as 4120.

CHROMIUM STEELS  
See Note (a) for Silicon Content

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus Max.	Sulphur Max.	Chromium Range
5120	0.15-0.25	0.30-0.60	0.040	0.050	0.60-0.90
5140	0.35-0.45	0.60-0.90	0.040	0.050	0.80-1.10
5150	0.45-0.55	0.60-0.90	0.040	0.050	0.80-1.10
52100	0.95-1.10	0.20-0.50	0.030	0.035	1.20-1.50

CHROMIUM-VANADIUM STEELS  
See Notes (a) and (c) for Silicon, Phosphorus, and Sulphur

S.A.E. No.	Carbon Range	Manganese Range	Chromium Range	VANADIUM	
				Minimum	Desired
6115	0.10-0.20	0.30-0.60	0.80-1.10	0.15	0.18
6120	0.15-0.25	0.30-0.60	0.80-1.10	0.15	0.18
6125	0.20-0.30	0.60-0.90	0.80-1.10	0.15	0.18
6130	0.25-0.35	0.60-0.90	0.80-1.10	0.15	0.18
6135	0.30-0.40	0.60-0.90	0.80-1.10	0.15	0.18
6140	0.35-0.45	0.60-0.90	0.80-1.10	0.15	0.18
6145	0.40-0.50	0.60-0.90	0.80-1.10	0.15	0.18
6150	0.45-0.55	0.60-0.90	0.80-1.10	0.15	0.18
6195(c)	0.90-1.05	0.20-0.45	0.80-1.10	0.15	0.18

TUNGSTEN STEELS  
Notes (a) and (d) for Silicon, Phosphorus, and Sulphur

No.	Carbon	Manganese	Chromium	Tungsten
71360	0.50-0.70	0.30 max	3.00-4.00	12.00-15.00
71660	0.50-0.70	0.30 max.	3.00-4.00	15.00-18.00
7260	0.50-0.70	0.30 max.	0.50-1.00	1.50- 2.00

SILICON-MANGANESE STEELS  
See Note (b) for Phosphorus and Sulphur

No.	Carbon	Manganese	Silicon
9255	0.50-0.60	0.60-0.90	1.80-2.20
9260	0.55-0.65	0.60-0.90	1.80-2.20

## 202 COMPARISON OF ALLOYING EFFECTS: S.A.E. STEELS

## CORROSION AND HEAT-RESISTING ALLOYS

See Note for Silicon, Phosphorus, and Sulphur Content

No.	Carbon	Manganese	Chromium	Nickel
30905	0.08 max.	0.20-0.70	17.00-20.00	8.00-10.00
30915	0.09-0.20	0.20-0.70	17.00-20.00	8.00-10.00
51210	0.12 max.	0.60 max.	11.50-13.00	.....
X51410	0.12 max.	0.60 max.	13.00-15.00	.....
51335	0.25-0.40	0.60 max.	12.00-14.00	.....
51510	0.12 max.	0.60 max.	14.00-16.00	.....
51710	0.12 max.	0.60 max.	16.00-18.00	.....

## NOTES TO TABLES OF CHEMICAL COMPOSITION

Silicon range of all S.A.E. basic open-hearth alloy steels shall be 0.15 to 0.30%. For acid alloy steels and acid open-hearth alloy steels, the silicon content shall be 0.15% maximum.

Note 1. Phosphorus and sulphur in all S.A.E. nickel steels, nickel-chromium steels, molybdenum steels, and silicon-manganese steels shall be 0.040% max. and 0.050% max., respectively.

Note 2. Phosphorus in all S.A.E. chromium-vanadium steels shall be 0.040% max. except in No. 6195, which shall be 0.039% max. Sulphur in all S.A.E. chromium-vanadium steels shall be 0.020% max. except in No. 6195, which shall be 0.035% max.

Note 3. Phosphorus and sulphur in all S.A.E. tungsten steels shall be 0.035% and 0.040% max. respectively.

Note 4. Silicon shall be 0.50% max. except in alloys 30905 and 30915, which may be 0.75%.

Phosphorus shall be 0.030% max. in all corrosion and heat-resisting alloys. Sulphur shall max. in all except in the free-cutting alloy X51410, which shall be in the range 0.15 to

**Customary Treatments for Some S.A.E. Steels.** The treatments suggested by the S.A.E. for some of the non-carburizing steels are summarized below. (The treatments for carburizing steels will be given in Chapter 14.) This is a summary only; further details are in the S.A.E. and A.S.M. Handbooks.

QUENCHING AND TEMPERING SUGGESTIONS OF THE S.A.E.  
STEEL TREATMENT

	Normalize From	Heat for Quenching to	Quench in	Draw*
1025	1700-1800°	1575-1650°	water	
X1025	"	"	"	
1030	"	"	"	
1035	1650-1750°	1525-1575°	water or oil	
1040	"	"	depending on section	
X1040	"	"	"	
1045	1600-1700°	1450-1550°	"	
X1045	"	1475-1525°	"	
1050	"	"	"	

		Heat for Quenching to	Quench in	Draw*
X1050	1600-1700°	1475-1525°	depending on section	
		1450-1550°	"	
X1055	Normalize or anneal to	1500-1550°	oil	
1060	desired structure	1450-1550°	"	
1065	"	"	"	
1070	"	"	"	
1075	"	"	"	
1080	"	"	"	
1085	"	1400-1500°	"	
1090	"	1430-1500°	oil, water, or brine	
1095	"	1500-1600°	oil (for leaf springs) and temper 750- 900°	
X1330 }	1650-1750°	1525-1575°	water or oil	
T1330 }				
X1335 }	1650-1750°	1500-1550°	"	
X1340 }				
T1335 }	1650-1750°	1525-1575°	oil	
T1340 }				
T1345 }	Norm. or ann. to desired structure	1450-1500°	"	
T1350 }		1450-1550°	"	
2330 }	"	1450-1500°	oil or water	
2335 }				
2340 }	"	1425-1475°	oil	
2345 }				
2350 }				
3125 }	1650-1750°	1500-1550°	oil or water	
3130 }				
3135 }	1650-1750°	"	oil	
3140 }				
X3140 }	To desired structure	"	"	(3150 for gears draw 375-425°)
3145 }				
3150 }				
3320 }				
3240 }	To desired structure	1475-1525°	"	
3245 }				
3250 }				
3325 }				
3335 }				
3340 }	"	1425-1475°	"	
3435 }				
3450 }	"	1400-1450°	"	
4130 }	"	1550-1650°	oil or water	
X4130 }				

		Heat for quenching to	Quench in	Draw*
4140	To desire	1525-1625° 1500-1600°	oil	
4940	"	1450-1500°	"	for gears heat for quenching in cyanide or activated bath, draw 350-450°
5120		1500-1650°†	oil or water	(see Chap. 14 for refinery treatments)
5140	Normalize or anneal to desired structure	1500-1600° 1475-1525°‡	oil "	(draw 375-425° for gears)
5150	"	1450-1550° 1475-1525°	" "	(draw 375-425° for gears)
52100	slow cool 7-35 hr. from 1350-1450° for spheroidizing	1425-1475° 1500-1600°	water oil	
6125 6130	norm. 1650-1750°	1575-1625°	oil or water	
6135 6140	norm. or ann. to desired structure	1550-1650°	oil	
6145 6150	"	1525-1625°	oil	
9255		1500-1650°‡	oil	
30905	annealing	Heat to 1850-1950°, hold 1 hr., quench in water, oil, or air (to produce austenitic structure)		
51210	annealing	1400-1450°, hold 1 hr., air or furnace cool		
X51410	hardening	1800-1850°, quench in oil, draw as desired		
51335	annealing	1425-1475°, hold 1 hr., furnace cool		
	hardening	1825-1875°, quench in oil or air, draw as desired		
51510	annealing	1475-1525°, hold 1 hr., air or furnace cool		
51710	annealing	1475-1525°, hold 1 hr., furnace cool		

\* In all cases, draw to required hardness.

† In cyanide or activated bath.

‡ Depending upon section.

"Microstructures of S.A.E. Steels After Recommended Heat Treatments," by Sauveur and Reed, published 1929 by E. F. Houghton & Co., Philadelphia, shows the structures, generally finely sorbitic, obtained by quenching as recommended and drawing at 1000° F. The hardness secured by the treatment is entered beneath the micrograph.

## BIBLIOGRAPHY

1. Bethlehem Steel Company: "Bethlehem Alloy and Special Steels." Bethlehem, Pa., The Author, 1935, 375 pages.
2. Ryerson & Son, Joseph T.: "Ryerson Certified Steels." Chicago, 1937, 26 pages.
3. BURNS, J. L., T. L. MOORE, and R. S. ARCHER: Quantitative Hardenability, *Am. Soc. Metals*, Preprint 5, October, 1937, 23 pages.
4. CAMPBELL, H. H.: The Influence of Carbon, Phosphorus, Manganese and Sulphur on the Tensile Strength of Open-hearth Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 35, 1905, pp. 772-810, discussion, pp. 1043-1048.
5. GROSSMANN, M. A.: "Principles of Heat Treatment." Cleveland, Am. Soc. Metals, 1935, 147 pages.
6. EPSTEIN, S., J. H. NEAD, and J. W. HALLEY: Choosing a Composition for Low-Alloy High-Strength Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 120, 1936, pp. 309-345. Also: *Metals Technology*, v. 3, 1936, Tech. Publ. 697, 31 pages.
7. MORRIS, M. J. R., R. SERGESON, and G. W. GABLE: Critical Temperatures for S. A. E. Steels, "National Metals Handbook," 1936 ed., pp. 361-362.
8. BAIN, E. C.: Some Characteristics Common to Carbon and Alloy Steels, *Yearbook Am. Iron Steel Inst.*, v. 24, 1934, pp. 86-128.
9. McQUAID, H. W.: The Importance of Aluminum Additions in Modern Commercial Steels, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 797-838.
10. McBRIDE, D. L., C. H. HERTY, JR., and R. F. MEHL: The Effect of Deoxidation on the Rate of Formation of Ferrite in Commercial Steels, *Ibid.*, v. 24, 1936, pp. 281-312.
11. KNOWLTON, H. B.: Physical Properties of Axle Shafts, *Ibid.*, v. 25, 1937, pp. 260-296.
12. JONES, J. A.: The Present Trend in Alloy Constructional Steels, *Iron Steel Ind.*, v. 10, 1936, pp. 194-201.
13. GREAVES, R. H., and S. H. REES: Mass Effect in Some Alloy Steels, *Metalurgist (Supp. to Engineer)*, February 25, 1938, pp. 103-108.
14. JANITZKY, E. J.: Influence of Mass in Heat Treatment, *Iron Age*, v. 110, 1922, pp. 788-790.
15. ESSER, H., W. EILENDER, and H. MAJERT: Der Einfluss verschiedener Legierungselemente auf die Abschreckhärbarkeit von Stahl (The Influence of Various Alloying Elements on the Hardenability of Steel), *Archiv Eisenhüttenw.*, v. 7, 1938, pp. 367-370.
16. KALLEN, H., and H. SCHRADER: Die Durchvergütung von Konstruktionsstählen unter Berücksichtigung des Einflusses von Stückquerschnitt und Legierung (The Deep Hardening of Structural Steels with Special Reference to the Influence of Cross-Section of the Piece and Composition), *Ibid.*, v. 4, 1931, pp. 383-392.
17. GILLET, H. W., and E. L. MACK: "Molybdenum, Cerium, and Related Alloy Steels." New York, The Chemical Catalog Co., 1925, 295 pages.
18. LESSELLS, J. M.: Notes on the Fatigue of Metals, *Mech. Eng.*, v. 45, 1923, pp. 695-696.
19. REED, E. L.: Influence of Special Elements on the Carbon Content of the Iron-Carbon Eutectoid, *Trans. Am. Soc. Steel Treat.*, v. 20, 1932, pp. 115-176.
20. GREGG, J. L., and B. N. DANILOFF: "The Alloys of Iron and Copper." New York, McGraw-Hill Book Co., 1934, 454 pages.

# COMPARISON OF ALLOYING EFFECTS: S.A.E. STEELS

21.           and F. POZEMIL: A Contribution on the Constitution of the Iron-Iron System. *J. Iron Steel Inst.*, v. 122, 1930, pp. 191-213.
22.           L.: Beitrag zum System Eisen-Kohlenstoff-Silizium (Contribution to the System Iron-carbon-silicon). *Stahl u. Eisen*, v. 50, 1930, pp. 1725-1727.
23.           WALTER, A. E., W. W. DELAMATER, and H. A. SCHWARTZ: The Composition and Critical Temperature of Pearlite Containing One Per Cent Silicon. *Am. Inst. Soc. Metals*, v. 22, 1934, pp. 120-138.
24. H. BELMONT, E.: "Einführung in die Sonderstahlkunde." (Special Steels.) Berlin, J. Springer, 1935, 566 pages.  
ZIMMERMAN, J. G., R. H. ABORN, and E. C. BAIN: Some Effects of Small Additions of Vanadium to Eutectoid Steel. *Trans. Am. Soc. Metals*, v. 25, 1937, pp. 755-757.
26. LAUDERDALE, R. H., and O. E. HARDER: Study of Carbide Solution in Hypoeutectoid Plain Carbon and Low Alloy Commercial Steels, Am. Soc. Metals, National Metal Congress, October 27, 1938, Paper presented.
27. BAIN, E. C.: Grain Size in Steels to be Heat-Treated, *Iron Steel Inst.*, Advance copy 2, October, 1938, 24 pages.
28. JOMINY, W. E.: Hardenability Tests, *Am. Soc. Metals*, Preprint 28, October, 1938, 23 pag
29.           W., and J. L. LAMONT: Hardenability of Low Chromium Steels, reprint 29, October, 1938, 19 pages.
30.           AS, G. T.: Transverse Hardness Tests of Heat Treated Steels, *Ibid.*, Preprint 30, October, 1938, 16
31. QUENEAU, B. R., and W. H. MAYO: Hardenability and Its Designation. The Hardenability Line, *Ibid.*, Preprint 31, October, 1938, 14 pages.
32. BURNS, J. L., and G. C. RIEGEL: Hardenability of Plain Carbon Steels, *Ibid.*, Preprint 32, October, 1938, 31 pages.
33. LUKESSEN, G. V.: Hardenability in Light Sections, *Ibid.*, Preprint 33, r. 1938, 16 pages.

## SECTION II. ENGINEERING STEELS OF MORE COMPLEX COMPOSITION

### CHAPTER 11

#### HIGH-YIELD STRENGTH, LOW-ALLOY STEELS

**Non-heat-treatable Steels.** In distinction to the ordinary requirements of an alloy steel in respect to ability to harden on quenching, there is a class of steel for which one of the most important requirements, aside from its strength, is that it shall *not* show a tendency toward hardening on rapid cooling, i.e., that it shall not only fail to show  $Ar''$  and become martensitic, but it shall also fail to produce such fine pearlite at  $Ar_1$  as would make it hard enough to impair its cold formability or give it any tendency toward brittleness.

**Weldability.** This requirement is imposed because these steels are in demand for use in making welded light-weight structures in the transportation industry. In welding, the metal of the weld, and that adjacent to it, are heated above the critical, and are cooled rapidly by the chilling effect of the parent metal which is being joined. The rate of cooling, when the parent metal is cold, may be of the order of that of an air blast, or even of an oil quench.

As a general guide, the American Welding Society<sup>1</sup> places a limit of 200 Brinell (96 Rockwell B) on the weld or any metal adjacent to it, in steel to be welded cold and without a stress-relief anneal. The steel maker stays within this limit by holding the C low, ordinarily below 0.15%, and selecting ferrite-forming elements that strengthen the steel in the as-rolled or normalized condition but that do not introduce appreciable tendency toward air hardening. If such elements are used, their amount is strictly limited.

Higher C grades are welded, but not, in the best practice, without a stress-relief anneal. The "American Welding Society Handbook" suggests that, if the metal adjacent to the weld shows any zone between 200 and 250 Brinell, slight preheat is desirable in heavy sections, and stress relief advisable for light and mandatory for heavy sections; if there is a zone of 250–300 Brinell, the preheat should be to



## HIGH-YIELD STRENGTH, LOW-ALLOY STEELS

at least 300°, and all welds should be stress-relief annealed; if the steel hardens to 350 Brinell, the preheat should be to 400° or higher and the weld should be immediately stress-relief annealed before it is allowed to cool.

In welded railway-car construction and in many other very large welded structures to which these steels are applied, complete stress-relief annealing of the completed structure is entirely impractical, and all that can be done is to apply local preheating and postheating to prevent or mitigate internal stresses. The attempt is seldom completely satisfactory. Hence the most important present-day group of these steels is made up of those which are thoroughly insensitive to rapid cooling after welding. Obviously, securing the desired strength by cold work, or by any type of heat treatment (by quenching and tempering, or by precipitation hardening) that would be undone by the welding heat, is difficult, if not impracticable. The strengthening is almost universally done by alloying.

**Steels for Riveting.** If the structure is to be built up by riveting instead of welding, the requirement of insensitivity to the welding heat falls away, and the desired high strength might be obtained by quenching and tempering, by precipitation hardening, by cold work, or by mere alloying. Actually, the pieces dealt with are generally large plates or structural shapes which are highly inconvenient to quench. Even heating for precipitation hardening would require equipment and time that would be justified only in special cases. Rolling cool, i.e., using a low finishing temperature, rolling the steel through, and somewhat below the critical, with spray cooling, is utilized in one commercial steel, but it would appear difficult to adjust this so that the properties of plates and shapes of different sections will meet the requirements. For these reasons, the steels for riveting, like those for welding, are usually given the desired properties by alloying.

**Properties Required. Yield Strength.** For the uses to which these steels are put, cold formability is ordinarily a prerequisite. The ductility and bend test requirements are therefore almost as severe as in ordinary C structural steel. Design is ordinarily based on yield strength; hence the desire is to step up the yield strength as much as is compatible with formability (actually, the yield strength is approximately doubled). In some cases, endurance limit may be a primary factor in design, so it may be desirable that the steel have relatively high endurance properties. The tensile strength naturally tends to rise as the yield and endurance are raised, but high tensile is not sought in itself, and may even be limited when such a limitation is thought to favor formability.

**Corrosion Resistance.** Since the weight saving made by the application of steels of high yield strength is achieved by reduction in section, it is necessary to avoid rapid corrosion or pitting which would make the life of the thinner section too short to be economical. This can be achieved either by proper painting or other protection, or by the selection of a composition which is itself more corrosion resistant to the service environment. Some of the high-yield-strength steels do have improved corrosion resistance to some environments, but it is necessary to remember that in other environments they may be no more resistant than ordinary C steel, and that, even in the environments to which they are resistant, the resistance is of a type that is built up only gradually. The steels *start* to rust as rapidly as any other steel, so, if appearance of bare metal or the staining of paint is a primary factor, their widely heralded "corrosion resistance" is not of value.

**Limitations of Carbon Steels.** Ordinary structural C steel of 0.20 C, 0.15 Si, 0.50% Mn (with or without 0.20% Cu), as-rolled, has a tensile of 50,000–60,000 and a yield of 25,000–30,000 with elongation in 8 in. of about 25–30%, and impact of around 25 to 35 ft.-lb. Charpy.

By raising the C to 0.40–0.45% the yield and tensile can be raised to 45,000 and 85,000 the elongation drops to 18–20%, and the impact resistance is at least halved. Decreased formability and the loss of ductility on cold working, as in bending and flanging, as well as air hardening and embrittlement in welding<sup>2</sup> or flame cutting,<sup>3,4</sup> prevented the acceptance of such a steel for use as-rolled. The hardness of a 0.45% C steel, originally 175 Brinell, may rise to 350 Brinell in rapid welding.

**Requirements.** In round figures, present requirements are, in the welding grade, freedom from hardening to more than 200 Brinell on welding, and for material about 0.75 in. thick, 50,000 yield, 20% elong. in 8 in., formability and bend test nearly as good as in ordinary structural (maximum tensile of 85,000 may be imposed in furtherance of formability), impact resistance nearly as high, and, in some cases, corrosion resistance twice as good as Cu-bearing structural. In the riveting grade, 55,000 yield, 20% elong. in 2 in. (equivalent to about 18% in 8 in.). The formability, bend, and impact may be slightly reduced from the welding grade. The corrosion resistance requirement is less often imposed for this grade, but, when it is, it is the same as in the welding grade. Actual specifications recognize that very thin plate will ordinarily show higher tensile and very heavy plate poorer ductility and bend. In very heavy plate a couple of thousand per square inch may be shaved off the yield-strength requirements. Actual properties of most of the material delivered usually show 2000 to 10,000

yield above the minimum requirement, but the designer normally figures his factor of safety on 50,000 yield for welding and 55,000 for riveting grades.

**Modifications in Design.** Since the modulus of elasticity remains the same as in ordinary structural steel, stiffness has to be obtained by alterations in design to compensate for the smaller section and produce the necessary section modulus.

**Alloying to Meet Yield Strength and Other Requirements. Manganese and Silicon.** The yield-strength requirements of the riveting grade have been long met in commercial steel by the use of Mn. Such steels ranged from 0.25 C, 1.75% Mn to 0.35 C, 0.75% Mn, with the Si usually increased, from the amount common in C structural, to about 0.30%; hence these steels became known as "silicon structural steels," a misnomer since the increased strength was primarily due to the increase in Mn.

Less frequently the Si was raised to 0.75–1.25%<sup>5</sup> and the Mn held at 0.60–0.90%. The steels of 0.25–0.35% C with Mn and Si as strengthening elements gave yield strengths of 45,000–65,000 lb. in.<sup>2</sup> and tensile strengths of 80,000–95,000 lb. in.<sup>2</sup> with elongations of perhaps 16–22% in 8 in. In the lower C steels the impact resistance was held at about the level of ordinary structural steel.

These steels found use in bridges and similar structures, and in shipbuilding, and were acceptable for riveted structures. They still survive for such uses, but are unsuitable for welding, and their ductility is rather low for satisfactory cold forming.

A grade intermediate in properties between the usual C structural and the present welding and riveting grade, which contains about 0.30 C, 0.20 or higher Si, and about 0.70% Mn, giving about 45,000 yield, 85,000 tensile, 17% elong. in 8 in., is covered by standard specifications (A.S.T.M.—A94 : 33) for "silicon structural steel," but present interest is more active in the steels of better yield and better ductility. The C-Mn steels are nowadays, in general, made with 0.25–0.30% C and 1–1.50% Mn to get a better combination of strength and ductility. Grain-control additions are also made to improve ductility and impact.

**Copper, Phosphorus, Nickel, and Chromium.** It had long been recognized that a small Cu content markedly increases the resistance to atmospheric corrosion and this was definitely proved by the A.S.T.M. tests.<sup>6</sup> As a result, at least 0.20% Cu is put into any steel of this class when specified. In this amount it has no perceptible effect on mechanical properties. The A.S.T.M. tests also provided strong indication that P, and especially its addition to the Cu steels, produced a like

benefit. This was pointed out in 1921 by Storey,<sup>7</sup> but there was quite a delay before it was generally realized and utilized. Nickel and Cr, when either is present in rather large amounts, are known to increase the atmospheric corrosion resistance, though their effects in such small quantities as are used in some of these steels are not so readily observable. The strengthening effect of Ni or Cr is, of course, well recognized.

**Corrosion Resistance.** Much emphasis is laid, by their producers, upon the corrosion resistance of these steels. For those in which more than 0.20% Cu (in the presence of high P) or appreciable amounts of Ni or Cr are used, the claims usually read that they have "4 to 6 times" the corrosion resistance of plain C steel, or "2 to 3 times" that of ordinary Cu-bearing steel.\* Such claims are based upon results of tests on bare sheet in atmospheric exposure, though in many cases the tests have not been of sufficiently long duration for final appraisal. As is the case with ordinary Cu-bearing steel, the resistance to rusting is due to the development, under suitable conditions of exposure, of a relatively impervious rust film whose presence slows down the later stages of rusting; and such a film is not formed unless suitable cycles of humidity and drying are provided by the environment. As Britton<sup>8</sup> shows, the steels *start* to rust just as quickly as C steel so that, when a paint film becomes pervious enough to allow the onset of rusting beneath it, the painted steel becomes unsightly and, in uses where the decorative effect is paramount, the steels have little if any advantage.

In submerged corrosion, or where opportunity for the rust film to dry out is not so great as in atmospheric corrosion, some of the steels may show a slight advantage over C steel, but none are truly resistant.

It is claimed<sup>9</sup> that steel No. 9 (Table I) with 0.40% Cu, around 0.13% P and 1% Cr, has such a balance of alloying elements as to be specially resistant to atmospheric corrosion when exposed as bare sheet, and the claim seems justified. However, several of the other steels, especially those high in P, rank very close to it. It is not yet possible to rate all these steels in the order of their corrosion resistance. Atmospheric corrosion resistance of bare sheet is a useful property but the impression should not be gained that these steels are entirely proof against corrosion, or even more usefully resistant than plain C steel unless the conditions of corrosion are favorable to them.

**Strengthening Effects.** The strengthening effect and the cheapness of Cu led to its use in larger amounts than in the older Cu-bearing

\* No special claim for corrosion resistance is made for steel No. 3 (Table I) in spite of its Cr content.

ds in which corrosion resistance only, and not strength, was sought, on resistance is increased as the Cu is raised from 0.20 toward 1.50%, but the increase above 0.20% is not so rapid as it is below that figure, while the strength supplied by Cu increases steadily and is more marked above 0.40% Cu.

To avoid the necessity of taking special care in the rolling of steels with more than 0.40% Cu, which are prone to edge crack, some Ni may be added,<sup>10</sup> since this addition combats that trouble. The Ni also exerts its regular strengthening effect and gives added atmospheric corrosion resistance to the steel. In the absence of Ni, soaking pit, reheating and rolling temperatures must be kept low. Steel with 1% Cu and no Ni has been so handled.<sup>11</sup>

When it became understood that, despite the embrittling effect of large amounts of P in medium-C and high-C steels, these low-C steels could utilize it as a ferrite strengthener without brittleness (see Chap 2), P was enrolled as an alloying element. It is of interest to note that some of these P-alloyed steels show phenomenally high endurance ratios (see Chap. 1, Vol. I) and that strength at high temperatures, a property especially bestowed by Mo, is also gained from P.<sup>12</sup>

**Grain Control.** The strengthening and grain-refining effects of a small amount of Mo, say 0.20% or more, or V, about 0.10%, are also utilized in some of these steels, especially in those carrying over 0.50% Mn. Since the steels are used in the as-rolled, or the normalized condition, an Al or Ti addition of about 0.05% to produce a high coarsening temperature and small grain will often be utilized in the steels whose composition would not otherwise produce them. Fine grain not only helps ductility and impact but it makes the steels less weld hardening, in line with the usual behavior of fine-grained, shallow-hardening steels.

**Endurance.** Several of the P-containing steels show remarkably high endurance limits in rotary bending (see Table I), the endurance limit approaching the yield strength and the endurance ratio rising to 70-75. Endurance ratios on higher C steels whose strength is gained from high Mn run around the usual 45-50, i.e., approximately half the tensile strength. Endurance data are lacking on most of the newer steels, and the relative propensities toward notch propagation, which have a bearing on actual performance in service under repeated stress as contrasted with a laboratory test on a specimen free from stress raisers, have yet to be evaluated.

**Available Elements.** With the various alloying elements listed above, as well as the older ones, Mn and Si, to choose from, it is ob-

vious that many different combinations of ferrite strengtheners might be used to attain the ends desired.

The ferrite-strengthening effects of the various alloying elements upon the low-C steels to be used as-rolled, without heat treatment, are discussed by Epstein, Nead, and Halley<sup>13</sup> whose plots for the effect of C, Cu, Mn, Ni, P, Si, Cr, and Mo in Fe or very low-C steels have been previously shown. From these the variations in strengthening effect are evident.

**Cost.** The cost of the alloying elements plays a considerable part in the selection of steel of this type. The riveting steels alloyed with Mn are usually low enough in Mn so that they do not take the "alloy extra," but, in the welding grade, other alloying elements that do take the extra are generally used.

It would be easy to meet the mechanical requirements by the use of Ni; a steel of 0.15 C, 0.15 Si, 0.30 Mn, 3.00% Ni may have a yield of 55,000 and a tensile of 75,000, with 25% elong. in 8 in., and up to about double the impact of ordinary low-C structural steel.

As Aston<sup>14</sup> points out, the whole group of other steels in this class can be described as an attempt to produce those properties more cheaply than with Ni. Clamer<sup>15</sup> long ago pointed out that some of the Ni could be replaced by Cu, and one of the present-day steels utilizes such a combination. In several of the steels the cheap alloying elements are supplemented by Mo or V, primarily for grain-size control, but, on the whole, most producers rely on cheap Mn, Si, Cr, Cu, and P in balanced amounts.

**Limitation of Carbon in the Welding Grade.** Bain and Llewellyn<sup>16</sup> set the upper limit of C content for good welding in the low-alloy steels as a class at 0.15 %, Kinzel<sup>17</sup> at 0.14 or 0.15%, Chapman<sup>18</sup> at 0.17%. Spraragen and Claussen<sup>19</sup> have examined the literature in an attempt to find out what combinations of C and Mn represent the upper limit for satisfactory welding, but it turned out that the literature deals with the subject chiefly from the point of view of being able to weld without the weld itself cracking. They were unable to determine the exact limit from published data, so the American Welding Society is considering a special research to find out. In view of the effect of grain size on the properties of normalized steels it is probable that the amount of C that can be borne in a welded steel of given Mn content will depend on the grain-coarsening characteristics. Manganese-V steels, for example, are thought to be weldable without brittleness, with somewhat higher C than the coarse-grained plain Mn steels. Some producers consider that 0.18 C, 1.35 Mn, 0.10% V is entirely safe for welding.<sup>20</sup> However, the American Welding Society found that a steel

of 0.17 C, 0.23 Si, 1.34 Mn, 0.15% V or one of 0.18 C, 2.5% Ni was weld hardening and required stress-relief annealing. Swinden<sup>22</sup> considers that a steel of 0.15 C, 1.44 Mn, 0.16 Si, 0.20% Cu is weld hardening and proposes to add Ti. An addition of 0.37% Ti to this steel, he found, would bind enough of the C to avoid weld hardening. The strength, with such an addition, was stated to be good, but exact figures were not given.

In general, 0.15% C is taken as the upper limit for any steel to be welded without preheating or stress-relief annealing. According to the "American Welding Society Handbook,"<sup>1</sup> a maximum tensile of \$5,000-85,000 should not be exceeded in these welding-grade steels, on the score of formability.

The purpose of the 0.15% max. C limit is to ensure that the metal in the zone adjacent to the weld shall not show any tendency toward air hardening, i.e., it must not go above 200 Brinell (96 Rockwell B). This is achieved in practically all the steels of this group so far put on the market, whatever the alloying elements, if the C does not go over 0.15%.

**Limitation of Phosphorus.** Phosphorus is a more potent strengthening element than C, but, unlike C, it does not tend toward weld hardening. These facts, coupled with its influence on corrosion resistance, make it especially fitted for use in the welding-grade steels. However, as has been shown in Chap. 2, if raised to too high a level, it has an embrittling effect on impact, even though static ductility is but slightly affected. The embrittling effect is very marked in the higher C steels. Although making the steel fine grained by Al, Ti, or V additions allows a higher P+C content than in a coarse-grained steel, and the presence of some other alloying elements, notably Cr, also raises the permissible P-C (while others, such as Si, seem to lower it), in any steel there will be some level of P+C which it is unsafe to approach.

Segregation of P is likely to occur, especially in large ingots, so that the sum of P-C must so be chosen that a dangerous level is not approached in the most segregated, uncropped part of the ingot. The tendency is not to let C rise much above 0.10% in the welding steels when P is over 0.10%. In general 0.25% P+C appears permissible. Some producers state the maxima of their ranges for P and C as running slightly above that sum, and some early heats exceeded the 0.25% P-C figure, but it is doubtful if steel of the welding grade is intentionally made today above that sum.\*

**Commercial Steels.** Leaving out the weaker "structural silicon"

\* For example, Steel No. 9 Table I, of recent production for best weldability is stated to contain 0.08 C, 0.12 P, 0.65 Si, 0.85 Cr, 0.30% Cu.

steel and the plain 3% Ni steel, representative ferrite-strengthened, low-C low-alloy, high-yield-strength (50,000 yield) steels for welding are listed in Table I. Table II shows representative 55,000-yield steels for riveting or other use not requiring welding or extreme cold formability. It should be noted that some of the steels of Table I show 55,000 yield and hence could equally well be placed in Table II. Some makers intentionally hold the welding grade at 50,000 yield, exceeding this as little as possible, in order to get maximum cold formability. At one time some makers advocated dropping the yield slightly below 50,000 on this score, but at present it appears that all aim to supply material of 50,000 minimum.

The tables are based on data collected by Cone,<sup>22</sup> supplemented and modified by later information from the literature and obtained by correspondence with producers. The data given usually represent average properties when these have been stated rather than specification minima, though some producers prefer to give only the latter, so that the tables do not necessarily rate each steel in correct relation to all the others. However, the tables do show that there is a variety of compositions that will give the properties required in each class.

The composition entered in the tables is an average composition, expected to give the properties shown. Each maker sets up his own ranges of composition, and the ranges generally vary according to the thickness of plate or size of section to be rolled, very large sections usually being given a trace more C and a few points more of the alloying elements. Several producers supply a variety of grades, the one shown for the welding grade usually being the softest one produced. For uses not involving welding or flame cutting, harder grades than those shown as the usual riveting grade are sometimes produced. Many of the grades intermediate between those shown in Tables I and II are discussed by Cone.<sup>22</sup>

**Other Steels.** The tables relate to American steels, commercially produced, and do not include all compositions that will meet the requirements as-rolled, or all the steels in commercial use. Compositions that might be added to Table I are two German steels, one of about 0.15 C, 0.45 Si, 1.15 Mn, 0.35 Cu, 0.20% Mo; the other about 0.15 C, 0.25 Si, 0.80 Mn, 0.55 Cu, 0.40% Cr, and to Table II, an English steel of about 0.22 C, 0.80 Mn, 0.30 Cu, 0.90% Cr.

**Evaluation of the High-yield-strength Steels.** The choice among these steels will ultimately be made on such bases as cost, uniformity of properties in plates of varying thickness, weldability, freedom from cracking on cold forming, resistance to impact at room and at low temperature, ease of handling in melting and rolling, and actual corro-



TABLE 1

REPRESENTATIVE COMPOSITIONS AND CORRESPONDING PROPERTIES. LOW-ALLOY STEELS FOR WELDING WITHOUT PRE-HEATING OR STRESS-RELIEVING ANNEALING. USUAL REQUIREMENTS, 50,000 MIN. YIELD, 21% ELONG. IN 8 IN. UP TO 1-IN. PLATE, LONGITUDINAL SPECIMENS

As rolled  $\frac{1}{2}$ -in. to  $\frac{3}{4}$ -in. plate or 1-in. rod unless noted.

No.	Approximate Composition, %										Tensile	Elong., % 8-in. 2-in.	R. A. %, 2-in.	Impact		Fu- durance Limit
	C	Si	Mn	P	Cu	Cr	Ni	Mo	V	Yield				Charpy	Izod	
1	0.13	0.15	0.40	.....	0.20 <sup>a</sup>	.....	.....	0.50 <sup>b</sup>	.....	50,000	80,000	23	55	.....	.....	.....
2	0.10 <sup>d</sup>	0.20	1.20	.....	0.20 <sup>a</sup>	.....	.....	.....	0.10	60,000	80,000	21	60-70	.....	60-110	.....
3	0.14	0.70	1.10	.....	0.20 <sup>a</sup>	0.50	.....	.....	.....	50,000	75,000	25	40	50	.....	.....
4	0.08	0.05	0.38	0.11	0.05	.....	0.75	0.10	.....	50,000	70,000	27	60	.....	130	48,000 (on sleeve)
5	0.09	0.40	0.70	0.10	0.60	0.25	0.33	.....	.....	50,000	75,000	25	.....	.....	75	49,000
6	0.10	0.50	0.75	.....	0.50	0.25	0.25	.....	.....	50,000	70,000	25	.....	.....	.....	.....
7	0.10	0.30	0.60	0.10	1.10	.....	0.55	.....	.....	50,000	75,000	25	60	55	.....	49,000
8	0.09	.....	0.70	.....	1.40	.....	0.80	0.10	.....	55,000	70,000	25	50	25	.....	45,000
9	0.10	0.75	0.20	0.13	0.40	0.90	.....	.....	.....	50,000	70,000	25	.....	40	60	45,000
10	0.08	0.30	0.55	.....	1.00	.....	2.00	.....	.....	55,000	75,000	25	60	40	.....	40,000
		max.														

<sup>a</sup> Cu content when Cu is desired.

<sup>b</sup> For heavy sections, 0.25% Mo for light sections.

<sup>c</sup> Atmospheric corrosion resistance greater than for Cu-bearing steel is claimed for steels 4-10.

<sup>d</sup> C content slightly above the usual level for this class, due to fine grain conferred by V. Manufacturers allow 0.18 C max.

Note that elements are given on 8-in. gauge lengths. Still later data collected by Case<sup>22</sup> show minor modifications in composition of some steels of Tables I and II, and slight changes in properties claimed. These groups of steels are still in a state of flux.

TABLE II

REPRESENTATIVE COMPOSITIONS AND CORRESPONDING PROPERTIES. NOT SPECIFICATION LIMITS. LOW-ALLOY STEELS FOR RIVETED STRUCTURES. SOME CAN BE WELDED WITH SUITABLE PREHEATING AND STRESS-RELIEF ANNEALING. USUAL REQUIREMENTS, 55,000 MIN. YIELD, 20% IN 2-IN., UP TO 1-IN. PLATE. LONGITUDINAL SPECIMENS.

As-rolled  $\frac{1}{2}$ - to  $\frac{3}{4}$ -in. plate unless noted.

No.	C	Si	Mn	P	Cu	Cr	Ni	Mo	V	Yield	Tensile	Elong. % 2 in.	R.A. %	Impact Izod
1	0.28	0.25	1.50	.....	0.50	.....	.....	.....	.....	55,000 to	80,000 to	20 to 35	.....	40 to 60
	0.35	0.15	1.40	.....	0.20	.....	.....	.....	.....	60,000	90,000	.....	.....	.....
2	0.30	0.50	0.90	.....	0.50	0.25	0.25	.....	.....	60,000 to	75,000 to	20	.....	.....
	max.	max.	max.	.....	.....	max.	max.	.....	.....	70,000	90,000	.....	.....	.....
3	0.20	0.20	1.45	.....	0.20 <sup>a</sup>	.....	.....	.....	0.12	65,000 to	90,000 to	20 to 30	60 to 70	.....
	.....	.....	.....	.....	.....	.....	.....	.....	.....	70,000	95,000	.....	.....	.....
4	0.17	.....	1.00	0.09	1.00	.....	.....	0.20	.....	55,000	85,000	20	.....	.....
	0.20	0.75	1.25	.....	0.20 <sup>a</sup>	0.50	.....	.....	.....	55,000	85,000	20 to 28	50 to 65	.....
6	0.25	.....	1.00	.....	1.40	.....	0.90	0.25	.....	70,000	90,000	15	.....	.....
	0.22	.....	0.80	.....	0.95	.....	1.90	.....	.....	65,000	90,000	25	55	.....
8*	0.13	.....	0.55	0.09	0.40	.....	.....	.....	.....	58,000	70,000	23	59	on $\frac{3}{4}$ -in. plate
	0.25	0.15	1.40	.....	0.25	0.12	.....	.....	.....	85,000	105,000	12	.....	.....

<sup>a</sup> When specified.

\* Rolled cold.

† Rolled cold.

Note that steels 2, 8, and 10 of Table I also show 55,000 yield.

Properties of steels 8 and 9 are for plate rolled below the critical. Other steels could similarly be strengthened and reduced in ductility. Since strength obtained by cold work would not be expected to be maintained in the conditions heated in welding steel, 8 has not been included in Table I, though the makers present data to show that the strength of welds is practically that of the plate. Not forming up to 1500° is said not to affect the properties. The endurance limit of 5 is given as 45,000.

Note that elongations are given on 2-in. gage lengths.

sion resistance in service in different environments. Sufficient data are not yet at hand to allow rating the various steels on these bases.

Since the steels require no heat treatment other than possibly a simple normalizing to even out small differences in properties in as-rolled plates of varying thickness, they will tend to supplant medium-C steels quenched and tempered to 50,000–55,000 yield.

Sufficient evidence as to usefulness is at hand so that the steels are coming into wide use, especially in the construction of light-weight freight cars. Street cars, trucks, and buses utilize them to some extent. Their use in shipbuilding is beginning.<sup>23, 24</sup>

Gibson<sup>25</sup> has described their use in heavy welded construction, such as turbine runners, engine bases in submarine service, and especially in such things as drag line buckets. When very large rolled shapes that now have to be built up by welding simpler sections are available, uses of this type will increase.

These steels seem to have little application to those automobile or aircraft needs which are ordinarily met by heat-treated steels, but, when the problem of increasing pay load comes up in construction so heavy that quenching and tempering are not applicable, the low-alloy steels deserve consideration.

This general class of steels has had much comment in the last few years, and descriptions of individual steels<sup>9, 11, 18, 26, 27, 28, 29</sup> as well as several summaries of the properties of the whole class<sup>22, 30, 31</sup> are available. Since the steels are not designed for heat treatment (indeed, those aimed for welding depend on their lack of response to heat treatment), they do not call for further attention here.

The purpose of this chapter has been to show the ferrite-strengthening effect of alloying elements and their interchangeability, matters that will come up for further attention as we deal with the simple and complex alloy steels designed for heat treatment.

#### BIBLIOGRAPHY

1. American Welding Society: "Welding Handbook." New York, The Society, 1938. 1210 pages.
2. WARNER, W. L.: Are Welding of Structural Alloy Steels, *Welding J.* (New York), v. 15, October, 1936, pp. 21–32 (suppl.).
3. GIBSON, A. E.: Nickel-Copper High-Strength Steels Must Have Specifications of Welded Construction, *Steel*, v. 100, January 25, 1937, pp. 34–39.
4. MOSS, H. H.: Structural Steels "Flame Softened" during Gas Cutting, *Iron Age*, v. 140, December 30, 1937, pp. 22–30. Also: Cutting and Tempering of Structural Steels, *Welding J.* (New York), v. 17, January, 1938, pp. 7–20.
5. GILBERT, H. W.: High Silicon Structural Steel, U. S. Bur. Stand., Tech. Paper 331, v. 21, 1926, pp. 121–143.
6. American Society for Testing Materials: Report of Committee A-5 on Corrosion of Iron and Steel, *Proc. Am. Soc. Test. Mat.*, v. 37, part 1, 1937, pp.

- 111-116. Results of 20-year exposure tests. See preceding volumes for complete tests. For Summary, see pp. 227-236 of GREGG, J. L., and B. N. DANIELLOFF: "The Alloys of Iron and Copper. New York, McGraw-Hill Book Co., 1934. 454 pages. Compare also, LORIG, C. H.: High Phosphorus Steels, *Metal Progress*, v. 33, July, 1938, pp. 60-61.
7. STOREY, O. W.: Discussion, *Trans. Am. Electrochem. Soc.*, v. 39, 1921, pp. 121-122.
  8. BRITTON, S. C.: Some Properties of Commercial Steel Sheets Containing Additions of Copper, Manganese, Chromium, and Phosphorus, *J. Iron Steel Inst.*, v. 135, 1937, pp. 161-185.
  9. WHETZEL, J. C.: Modern Steels and Weight Reduction, *Yearbook Am. Iron Steel Inst.*, v. 25, 1935, pp. 104-140.  
BRUNNER, J. W.: High Tensile Steels for Structural Purposes, *J. West. Soc. Eng.*, v. 40, 1935, pp. 103-108.
  - SCHRAMM, G. N., E. S. TAYLORSON, and A. F. STUEBING: New Alloy Steels and Their Application to Car Equipment, *Iron Age*, v. 134, December 6, 1934, pp. 33-38.
  - STUEBING, A. F.: High Tensile Steels, *Ry. Mech. Engr.*, v. 111, November, 1937, pp. 517-520.
  10. LORIG, C. H.: Discussion, *Trans. Am. Inst. Min. Met. Eng.*, v., 105, 1933, pp. 165-167.
  11. Ore Car Weighs Sixteen Per Cent Less Than Previous Design, *Railway Age*, v. 103, October 30, 1937, pp. 599-600.
  12. CROSS, H. C., and D. E. KRAUSE: Phosphorus as an Alloying Element in Steels for Use at Elevated Temperatures, *Metals & Alloys*, v. 8, 1937, pp. 53-58.
  13. EPSTEIN, S., J. H. NEAD, and J. W. HALLEY: Choosing a Composition for Low Alloy High Strength Steel, *Trans. Am. Inst. Min. Met. Eng.*, v. 120, 1936, pp. 309-345. Also: *Metals Technology*, v. 3, 1936, Tech Publ. 697, 31 pages.
  14. ASTON, J.: Corrosion in Relation to Engineering Structures, *Proc. Am. Soc. Civil Eng.*, v. 62, 1936, pp. 1239-1249.
  15. CLAMER, G. H.: Cupro-Nickel Steel, *Proc. Am. Soc. Test. Mat.*, v. 10, 1910, pp. 267-279.
  16. BAIN, E. C., and F. T. LLEWELLYN: Low-Alloy Structural Steels, *Proc. Am. Soc. Civil Eng.*, v. 62, 1936, pp. 1184-1200.
  17. KINZEL, A. B., and W. CRAFTS: "The Alloys of Iron and Chromium." New York, McGraw-Hill Book Co., Vol. 1, 1937, 535 pages.
  18. CHAPMAN, E.: Welded Structures, *J. Am. Soc. Naval Eng.*, v. 48, 1936, pp. 476-483.
  19. SPRAGEN, W., and G. E. CLAUSSEN: The Effect of Total Carbon and Manganese on the Mechanical Properties and Structure of Welded Joints in Plain Low-carbon Steel, A Review of the Literature to January 1, 1937. *Welding J.*, (New York), v. 16, September, 1937, pp. 23-32 (suppl.).
  20. CRITCHETT, J. H.: Report of Subcommittee on Welding of Low-Alloy Steel. *Ibid.*, v. 17, January, 1938, pp. 8-14 (Research Suppl.).
  21. SWINDEN, T.: Special Steels and Their Application to Engineering and Shipbuilding, *Metallurgia*, v. 17, 1938, pp. 181-185.
  22. CONE, E. F.: Carbon and Low-Alloy Steels, Am. Soc. Test. Mat., "Symposium on High-Strength Constructional Metals." Philadelphia, March 4, 1936, pp. 1-30.

23. TRASK, E. P.: Proposed Use of Alloys in Merchant Shipbuilding, *Metals Technology*, v. 3, October, 1936, pp. 66-83.
24. McALLISTER, L. P.: Modern Materials for High Pressure Boiler and Pressure Vessels, *Iron Age*, v. 135, June 6, 1935, pp. 24-26, 108, 110, 112.
25. GIBSON, A. E.: The Development and Use of Low Alloy High Tensile Steels, *Welding J.* (New York), v. 14, September, 1935, pp. 2-8.
26. KINZEL, A. B.: Silicon-Manganese Steels with Chromium Additions for Engineering Applications, *Trans. Am. Soc. Steel Treat.*, v. 14, 1928, pp. 866-876. A Low Alloy Steel for Large Tonnage Applications, *Iron Age*, v. 128, 1931, pp. 1686-1688.
- MILLER, W. B.: Properties of Cromansil Steels for Structural Purposes, *Steel*, v. 94, May 21, 1934, pp. 32-34, 35, 50.
27. LAWRENCE, V. H.: Engineering Aspect of High-Strength Sheet Steels, *Steel*, v. 96, January 21, 1935, pp. 27-29.
- LAWRENCE, V. H.: High Strength Corrosion Resistant Ferrous Alloy Sheets, U. S. Patent 2,111,454, March 15, 1938.
- LAWRENCE, V. H.: Process of Producing Sheets from Corrosion-Resistant Ferrous-Base Alloys Combining Improved Strength Properties with Corrosion Resistance, U. S. Patent 2,108,588, February 15, 1938.
- Booklet, "A.W. Presents Dyn-el," April, 1938. Alan Wood Steel Co., 43 pages.
28. MILLER, H. L.: Alloy Steel (Copper-Nickel-Molybdenum), *Metal Progress*, v. 28, July, 1935, pp. 28-31, 70.
- MILLER, H. L. and T. R. LICHTENWALTER: Using High-strength Low-alloy Steel, *Steel*, v. 101, October 15, 1937, pp. 62-63.
29. DOUGHERTY, R. S. A.: New Low Alloy Steels with Unusual Properties Are Results of Research, *Ibid.*, v. 101, September 13, 1937, pp. 36-39.
- DOUGHERTY, R. S. A.: Manari R (A Corrosion Resistant, High Strength Steel), *Iron Age*, v. 140, September 9, 1937, pp. 44-48.
30. GILLET, H. W.: Trends in the Metallurgy of Low-Alloy High-Yield-Strength Structural Steels, *Metals Technology*, v. 3, October, 1936, pp. 40-61.
31. Low Alloy, High Yield-Strength Structural Steels (An extended abstract), *Metals & Alloys*, v. 7, 1936, pp. 77-84.
32. CENE, E. F.: The Present Status of the Low Alloy High Strength Steels, *Metals & Alloys*, v. 9, 1938, pp. 243-254.

## CHAPTER 12

### COMPLEX ALLOY STEELS BASED ON FERRITE FORMERS LARGE SECTIONS. THE NICKEL-CHROMIUM AND SIMILAR FAMILIES

**The Problem of Strength in Large Sections.** The vast majority of parts made from heat-treated steels are of relatively small cross section. For this reason the property charts, ordinarily shown to bring out the effects of an alloying element in a quenched and tempered steel, or the effects of variation in quenching temperature, quenching medium, and draw temperature, are based on results obtained in the treatment of bars of  $1\frac{1}{8}$ -in., 1-in., or even smaller diameter.

However, important objects such as die blocks, big guns, large shafts, huge gears, and load-carrying parts of heavy machinery often require high physical properties even at the center of large masses. It is possible to secure nearly as high physical properties at the center as at the surface of sections up to around 4-in. diam. by proper choice of alloying elements that confer depth-hardening properties. Above some such limit of dimensions, the effectiveness of martensitic hardening decreases so rapidly that, instead of an effort to harden completely, the effort becomes one of securing fine pearlite by normalizing. In either case, the critical cooling rate is the fundamental matter to be considered.

**Critical Cooling Rate.** In order to produce martensite (and by tempering it later, secondary troostite or sorbite), a steel must be cooled rapidly enough so that its critical cooling rate is exceeded, so that  $Ar'$  may be avoided and  $Ar''$  reached. Similarly, in the production of primary troostite and emulsified pearlite a rapid cooling rate is necessary, since a slow one produces coarse pearlite. The cooling rate is slightly increased by increase in quenching temperature, but, on the whole, attempts to speed up the cooling rate are severely limited by the low thermal conductivity of steel. It takes a long time to drive heat to the center of a large steel section in heating for forging, quenching, or normalizing, as has been brought out in many parts of Vol. I.

**Limitations of Heat Flow.** The abstraction of heat *from* the center of a large section is also a lengthy process. The most drastic quenching will pull the heat out of only a thin shell on the surface, while the interior retains it for a long time. Material of higher thermal conductivity, like Cu, or of lower density, like Al, will transmit heat from center to outside more rapidly than steel, which has both low thermal conductivity and high density. But the centers of large pieces even of Cu or Al do not cool rapidly enough to reach the cooling rate that steel requires to be hardened. Large mass defeats even higher conductivity than it is possible to reach in steel.

Thus, there is little that can be done to hasten the center cooling rate of large sections; the situation must be accepted as nature has made it.

**Alloying, the Only Recourse.** Something can be done in altering the nature of the steel so that it will harden even though it cools slowly. That is, the critical cooling rate may be decreased by making the steel coarse grained instead of fine grained, if this is permissible on the score of impact, and still more effectively, by alloying. As Fig. 9, Chap. 6, Vol. I shows, increase in C decreases the critical cooling rate, but only to a limited degree; and when C is increased too far the steel lacks ductility. Addition of alloying elements is the only way that the critical cooling rate can be so decreased that the attainable cooling rate at the center of a large section will suffice to harden the steel. Even this means fails in extremely large sections. Mochel<sup>1</sup> describes a turbine rotor forging 5 ft. diam. by 20 ft. long, weighing 55 tons. No attempt is made to harden the center of such a forging. Even with a highly alloyed steel the attainable strength at the center is only about 100,000 tensile, the center properties being practically those of an annealed steel, since the center would cool at an annealing rate no matter how drastically the outside were quenched. The best that can be done is to adjust the composition so that on normalizing (which is at a much slower cooling rate than normalizing of small sections) some approach toward fine pearlite may be had. Adjustment of composition to size is the key to obtaining optimum properties.

By alloying to produce what is, in small sections, an air-hardening steel, the critical cooling rate can be lowered so that, on oil quenching or air cooling, martensitic hardening can be obtained for some distance beneath the surface, and rather fine pearlite at the center of about a 10-in. diam. section.

**Characteristic Compositions.** Experience has led to the use of deep hardening steels of compositions like the following: for large die blocks, large shafts, big guns, and the like:

% C	% Mn	% Ni	% Cr	% Mo	% V	% W	% Cu
0.25	0.50	3.00	1.50	0.50	0.20	....	....
0.35	0.70	3.00	0.80	0.40	....	....	....
0.35	0.50	4.25	1.25	....	....	0.75	....
0.30	0.40	4.00	1.35	....	....	....	....
0.32	0.70	1.75	0.75	0.50	0.15	....	....
0.33	0.60	3.50	1.00	....	....	....	....
0.35	0.50	3.00	0.80	....	....	....	....
0.35	0.50	3.50	1.50	....	....	....	....
0.40	0.50	2.50	0.80	0.75	....	....	....
0.45	0.50	2.00	1.00	0.35	....	....	....
0.50	0.75	1.00	0.75	0.75	....	....	2.00

Prominent among available steels are, of course, those of the S.A.E. series:

S.A.E. No.	C %	Mn %	Ni %	Cr %	Mo %
3240	0.35-0.45	0.30-0.60	1.50-2.00	0.90-1.25	....
3335	0.30-0.40	"	3.25-3.75	1.25-1.75	....
3435	0.30-0.40	"	2.75-3.25	0.60-0.95	....
4340	0.35-0.45	0.50-0.80	1.50-2.00	0.50-0.80	0.30-0.40
4345	0.40-0.50	"	"	0.60-0.90	0.15-0.25

**The Mass Effect.** The mass effect in a variety of quenched alloy steels has been shown in Fig. 12, Chap. 10, where 0.75-, 2-, 4-, and 8-in. diam. specimens are compared on the basis of hardness at the outside. The only steels in that figure that hardened above 400 Brinell at the outside on oil quenching in 8-in. sections had the following analyses:

% C	Mn	Cr	W
0.32	4.16	1.57	
0.31	4.10	1.16	0.97

Gregg<sup>2</sup> cites data for air-quenched and tempered steels, and "Vanadium Steels and Irons"<sup>3</sup> gives data for oil-quenched and tempered steels, in large sections, the specimens being taken midway between center and edge:



% C	Ni	Cr	Mo	V	Diam. in.	Tensile	Yield	Elong. %	R.A. %
0.30	3.00	1.50	0.50	0.20	10-15	115,000	95,000	18	52 air <sup>2</sup>
0.55	1.50	0.75	0.30	.....	17	120,500	90,500	20	42 air <sup>2</sup>
0.30	1.79	0.70	0.46	0.11	17	98,000	70,000	17-27	40-65 oil <sup>2</sup>
0.33	1.85	0.74	0.41	0.11	12	120,000	95,000- 100,000	17-21	40-55 oil <sup>2</sup>
0.33	1.85	0.74	0.41	0.11	12	120,000	90,000- 95,000	18	40 air <sup>2</sup>

The last two lines show how little difference is made by oil quenching over air cooling.

**Reasons for Selecting the Nickel-Chromium Base for Large Sections.** The compositions cited above indicate the type of alloying additions that are made, that is, in steels for high strength in large sections. Something between 1.40 and 4% Ni, a ferrite-strengthening element, plus 1 to 2.25% total of Cr, Mo, W, or V, all carbide-forming elements, is commonly used, with C between 0.30 and 0.40%. If we specify a certain degree of toughness, e.g., if we temper the oil-quenched steel so that it shows 22% elong. the effect of building up the alloy content may be seen from the following data, collected from various sources.

TABLE I  
COMPOSITION, YIELD STRENGTH, AND REDUCTION OF AREA OF STEELS TREATED TO GIVE ELONGATION OF 22% AT CENTER OF 4-IN. DIAM. SECTIONS

Steel Composition, %					Yield Strength		R.A. %
	Ni	Cr					
0.35		1.00				65,000	65
0.40	3.50					70,000	55
0.40	1.50	0.60				70,000	55
0.40	1.75	1.00				75,000	60
0.30	2.20	1.10				83,000	65
0.25	2.50	0.60	0.50			85,000	65
0.35	.....	1.10	0.20			90,000	60
0.40	3.50	0.85				90,000	60
0.40	3.50	1.50				95,000	60
0.32	1.75	0.70	0.45		0.15	115,000	63
0.31	4.10	1.20	.....	1.00		128,000	65
0.40*	2.50	0.75	0.75			134,000†	60
0.30*	3.00	1.35	0.55		0.13	125,000‡	60

These steels normally carry from 0.30 to 0.80% Mn.

\* Composition approximate.

† 330 Brinell, 20% Elong., 50 ft.-lb. Isod.

**Nickel Compared with Manganese.** One might wonder, since Mn is both a ferrite strengthener and a carbide former, why large amounts of Mn are not used instead of the Ni-Cr combination. The limitations of high-Mn steels have already been discussed in Chap. 3, though the benefit of 1.00–1.25% Mn in large forgings as compared to C steel of usual Mn content was also there brought out. Strauss<sup>4</sup> has made comparisons of plain Mn and plain Ni in large sections, shown in

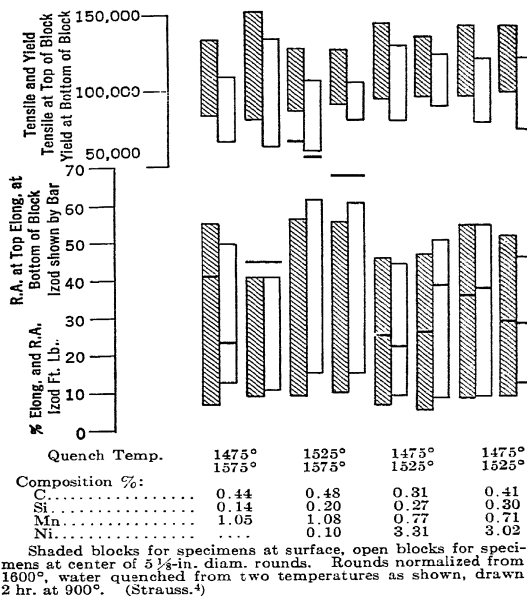


Fig. 1.—Comparison of 1% Mn and 3% Ni in heavy sections.

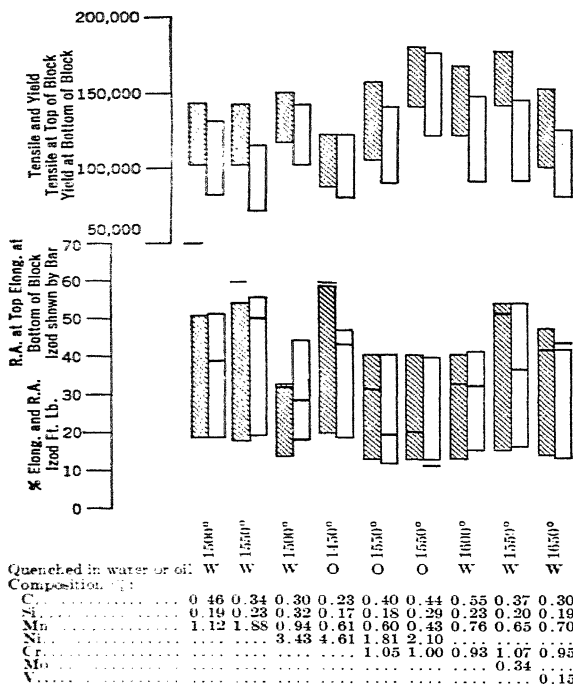
Fig. 1, and of a variety of steels shown in Fig. 2 (in these, yield was reported as proportional limit). The latter again indicates that the Ni-Cr combination with upwards of 2% Ni is outstanding for high yield strength.

Replacement of Ni by Mn in cast steel is common, as has been shown in Chap. 3. Greaves<sup>5</sup> studied the partial substitution of Ni by Mn in wrought steels in 2.25-in. by 3.25-in. specimens. After oil quenching from 1650° and drawing at 1110°, to give 25–28% elong., 61–64% R.A., his steels showed:

# COMPLEX ALLOY STEELS BASED ON FERRITE FORMERS

C %	Mn %	Ni %	Tensile	Yield
0.33	1.04	1.45	95,500	71,500
0.32	1.08	1.95	106,000	82,000
0.30	1.32	1.49	103,000	76,000
0.29	1.27	2.06	108,000	84,000
0.30	1.10	1.00	108,000	73,500
0.38	1.07	1.37	107,500	72,000
0.37	1.03	1.95	107,000	72,500

He concludes that in large sections "it may be necessary to reduce the C to 0.30% and increase the Ni to 1.5 or 2.0%." He based this



Shaded blocks for specimens at surface, open blocks for specimens at center of 3 1/2-in. diam. rounds. Rounds annealed at 1450-1500°, then quenched as shown, drawn 2 hr. at 900°. (Strauss.)

FIG. 2.—Comparison of depth hardening in Mn, Ni, NiCr, Cr, CrMo and CrV steels.

conclusion primarily on the impact values, since the 0.40% steels of 1% Mn and below 1.50% Ni gave impact below 40 ft.-lb., and the 0.29 C, 1.27 Mn, 2.06% Ni gave 59. Only partial substitution of Ni by Mn seems feasible, when depth hardening is sought, on the basis of these experiments, though the question of partial substitution in a Cr-containing steel is not answered thereby.

**Replacement of Nickel by Copper.** Partial substitution of Ni by Cu in large die blocks in which deep hardening is required is stated by Finkl<sup>6</sup> to be feasible.

**Nickel-free Steels.** Steels without strong ferrite formers, i.e., Cr, Mo, or Cr-Mo-V compositions, have depth-hardening characteristics and have been used, especially in Germany, for large sections; but (see Chap. 13) a rather critical draw temperature is required to reach the desirable balance between yield strength and ductility. Toughened ferrite, such as is produced by Ni, either alone or with some of its normal content replaced by Mn or Cu, seems needed to secure optimum properties in very large sections, and to be the less readily dispensed with as the sections increase up to the limit of size in which oil quenching is effective.

Gregg<sup>2</sup> cites data from Watertown Arsenal on 3- by 4-in. specimens, oil quenched from 1560°, drawn at 1200°. They are not entirely consistent in that B and C show strength in reverse order of C content, perhaps because of the higher Mo in B.

	% C	% Mn	% Ni	% Cr	% Mo	Tensile	Yield	Elong. %	R.A. %
A.....	0.30	0.44	4.51	.....	0.47	137,000	110,000	23.5	62
B.....	0.33	0.75	3.37	0.59	0.38	130,000	104,000	24	62.5
C.....	0.38	0.68	3.35	0.84	0.31	119,500	86,000	25	65
D.....	0.38	0.68	3.31	1.02	0.35	127,500	100,000	24.5	65

**Chromium and Molybdenum in Nickel Steel.** The reason for the introduction of Cr along with Ni into steels for large sections is evident from Table I and Fig. 2. Medium Ni plus around 1% Cr produces higher yield at a given ductility than is obtained with higher Ni and gives a cheaper alloy.

The choice of the relatively expensive Mo to replace part of the cheap Cr, or to augment it, is on technical grounds.

**Temper Brittleness.** As has been mentioned under precipitation hardening, steels with even moderate amounts of Cr or Mn, and such steels plus Ni, are likely to become brittle if slowly cooled from the

draw. The higher Cr, Ni-Cr steels are very likely to be temper brittle especially at the centers of large sections, which do cool slowly. The phenomenon is most commonly spoken of in relation to such steels. The Mn steels with 1 to 2% Mn, and even the plain 3.5% Ni steel, may likewise show it.

The addition of 0.30–0.50% Mo prevents the embrittlement. Jones<sup>7</sup> reports the following on steels oil quenched from 1650° and drawn at 1200°.

Ni	% Cr	% Mo	Izod Impact Ft.-lb. Quenched in Water from 1200°	Very Slow-cooled from 1200°
0.23	3.72	0.92	44	6
0.25	2.84	0.85	37	35

About 0.30% Mo was shown to be sufficient to remove temper brittleness in such tests. Still smaller amounts are effective in many steels.

A similar embrittlement occurs in some steels in long high temperature service, e.g., a year at 800°, and in such service larger amounts of Mo, i.e., 0.50% or more, are employed to safeguard against it. Some authorities think that Mo practically removes temper brittleness, others find that embrittlement may still occur, but all agree that Mo at least greatly delays its appearance.

**Machinability.** Another reason for the use of Mo is machinability. It is generally recognized that it is possible to machine the steels containing Mo at hardnesses somewhat above the 270 Brinell, 135,000 tensile often being taken as the limit of practical machinability. The explanation usually given is that the high reduction of area characteristic of such Mo steels at 140,000–160,000 tensile localizes the stressed volume of steel at the tip of the cut. (Compare Fig. 13, Chap. 2, Vol. I.) Boegehold<sup>8</sup> says that production machining of the Mo steels can sometimes be done at 440 Brinell, 220,000 tensile or even a trifle higher; that is, it may not be necessary to resort to grinding, though naturally, machining would not be extremely rapid.

**Vanadium in Nickel-Chromium and Nickel-Chromium-Molybdenum.** The addition of V, in amount usually ranging from 0.12 to 0.20%, or higher, is made to Ni-Cr or Ni-Cr-Mo steels, primarily for grain-refining purposes. A perceptible alloying effect is also exerted in this range of V. Abram,<sup>9</sup> working with 1½- by ¾-in. specimens, oil quenched from 1650° and drawn at 840–1290°, studied 13 steels, all with 2.05–2.80% Ni and 0.65–1.05% Cr. To these base

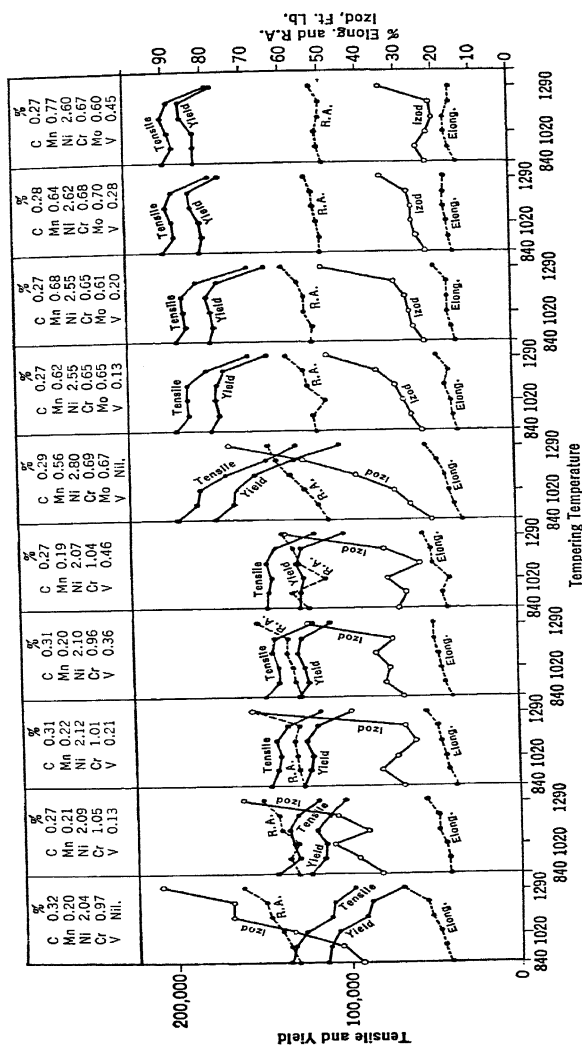
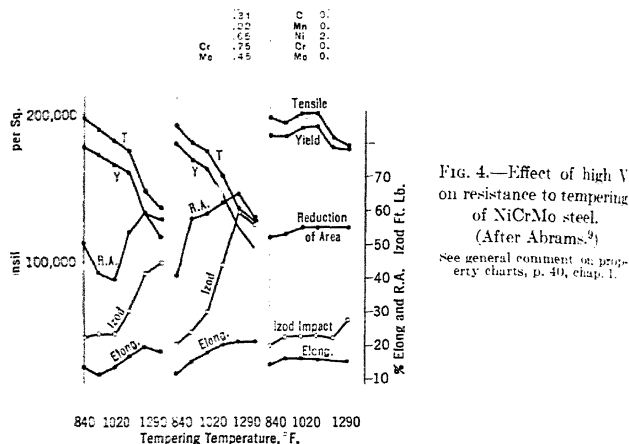


Fig. 3.—Effect of Mo and V in NiCr steel. (After Abrams.<sup>2</sup>) Specimens,  $1\frac{1}{2} \times \frac{3}{4}$  in., oil quenched from 1650°, drawn as shown. See general comment on properly charts, p. 40, chap. 1.

steels additions of 0.50–0.70% Mo or 0.13–0.46% V, or both, were made. In order to reach 200,000 tensile, within this range of draw temperatures and in this size, both Mo and V were required, as Figs. 3 and 4 show. The maintenance of high tensile and yield strength at



high draw temperatures by the combined addition of Mo and V is worthy of note.

**Total Alloy Content.** To obtain the maximum strength attainable in quenched and tempered steel, along with the toughness represented by 15% elong., 50% R.A., and 20 ft.-lb. Izod, it appears necessary to use about 4% alloying elements in a steel of about 0.30% C. This 4% is made up of about 2.50 Ni, 0.75 Cr., 0.50 Mo, 0.25% V. The Ni may vary from 2.00 to 3.00% and the Cr from 0.75 to 1.40%, Mo from 0.20 to 0.75% and V from nil to 0.50%, and the total of these alloying elements may range from 3 to 5%.

Probably because of the cost, the large quenched and tempered objects, like large die blocks, in which maximum strength and greatest depth hardening are sought, within economical limits, tend to be made from steel with a little higher C, lower Ni, higher Cr, lower Mo, and with the use of a cheaper grain refining addition than V.

**Range of Properties.** Depending on the draw temperature, properties ranging from something over a tensile of 250,000 and 200,000 yield, with over 10% elong., to 150,000 tensile, 100,000 yield, and 25% elong., can be secured from steels of this class in small wrought

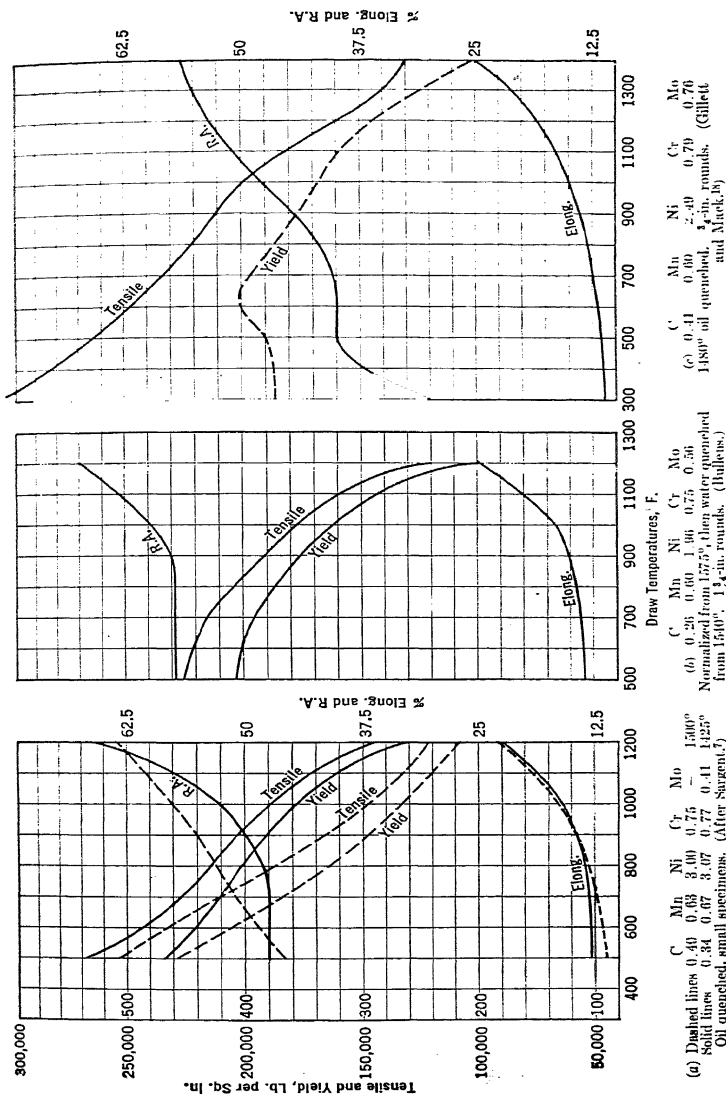


Fig. 5.—Properties of NiCrMo steels of relatively high Mo content, water and oil quenched, drawn as shown. See general comment on property charts, p. 40, chap. I.



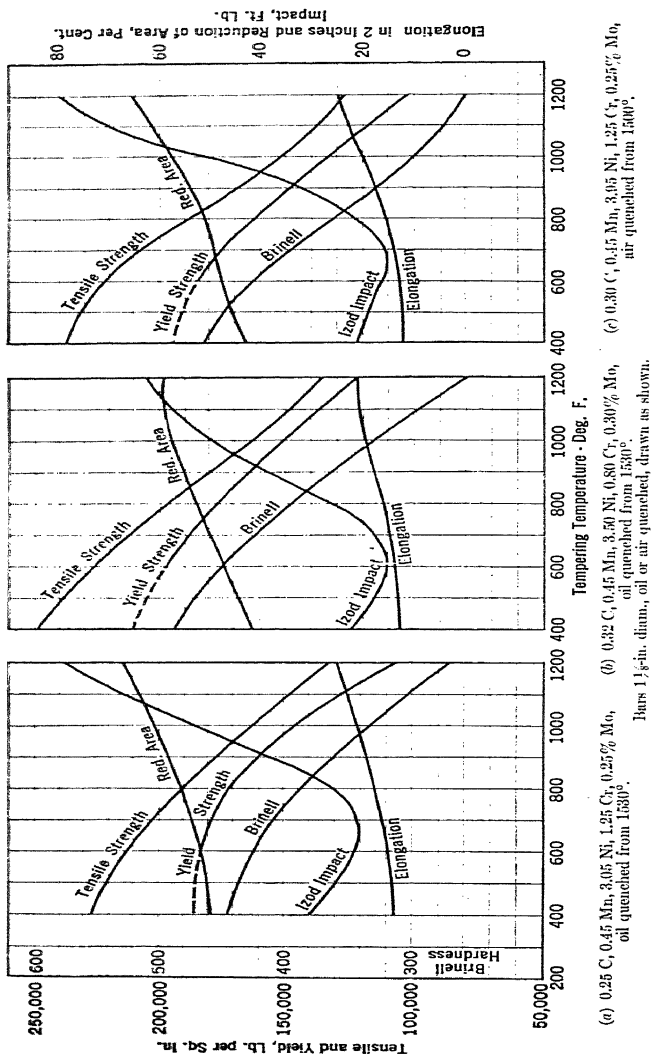
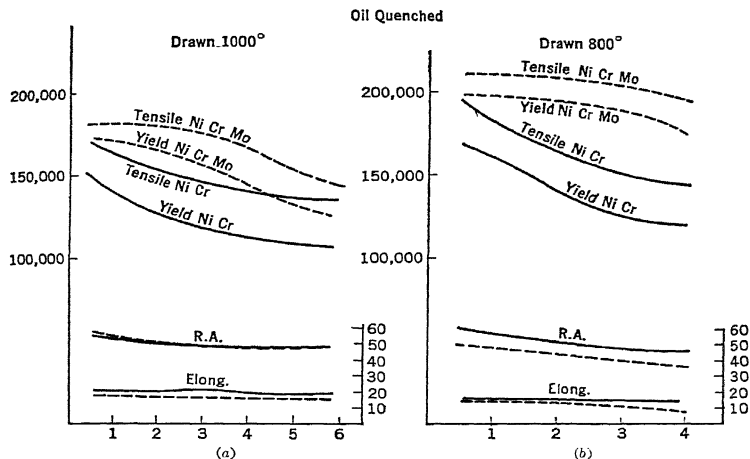


Fig. 6.—Properties of NiCrMo steels. (Nickel Alloy Steels, 12)

sections. Castings that are of such size and shape as may be quenched, made of similar steels but in a lower C range, also show very high properties in small sections. Property curves for small sections of Ni-Cr-Mo steels are included in Figs. 3 and 4, and others are shown in Figs. 5 and 6. Detailed discussions of these steels are given by Gregg<sup>2</sup> and Kinzel and Crafts.<sup>10</sup>



- (a) Solid lines, Ni Cr Mo; 0.45 C, 0.60 Mn, 1.80 Ni, 0.75 Cr, 0.25% Mo, oil quenched up to 2 in. from 1525°; 2 to 4 in., 1550°; over 4 in., 1575°, drawn at 1000°. Dashed lines, S.A.E. 3240; 0.40 C; 0.45 Mn, 1.75 Ni, 1.10% Cr, oil quenched, up to 2 in., from 1475°; 2 to 4 in., 1500°; over 4 in., 1525°, drawn at 1000°.
- (b) Solid lines Ni Cr Mo, as in (a), but drawn at 800°. Dashed lines, S.A.E. 3435, 0.35 C, 0.45 Mn, 3.00 Ni, 0.75 Cr, oil quenched, up to 2 in. from 1425°, over 2 in., from 1450°, drawn at 800°.

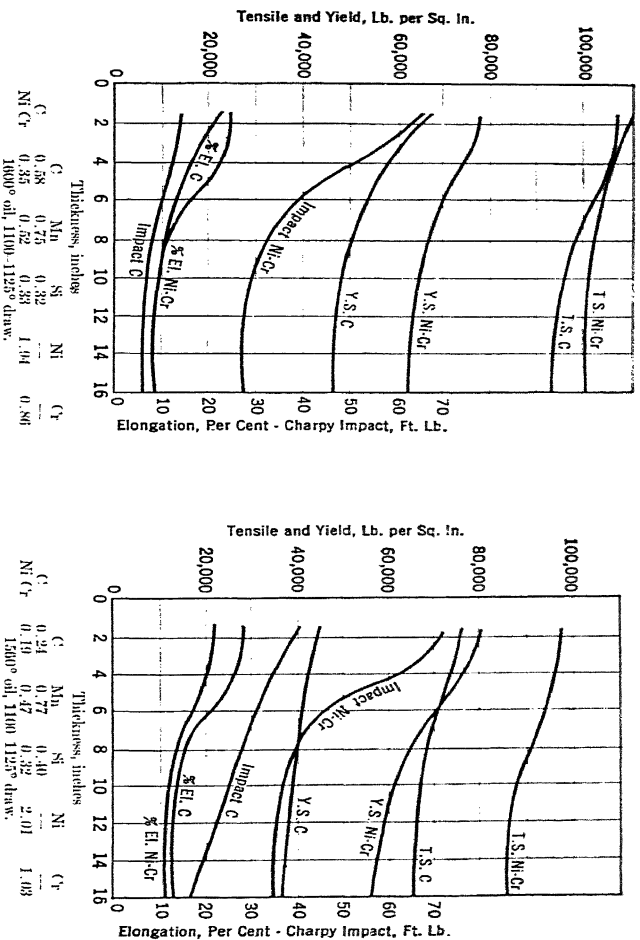
Fig. 7.—Size effects in quenched and drawn steels. Longitudinal specimens, cut midway between axis and surface. (Nickel Alloy Steels.)

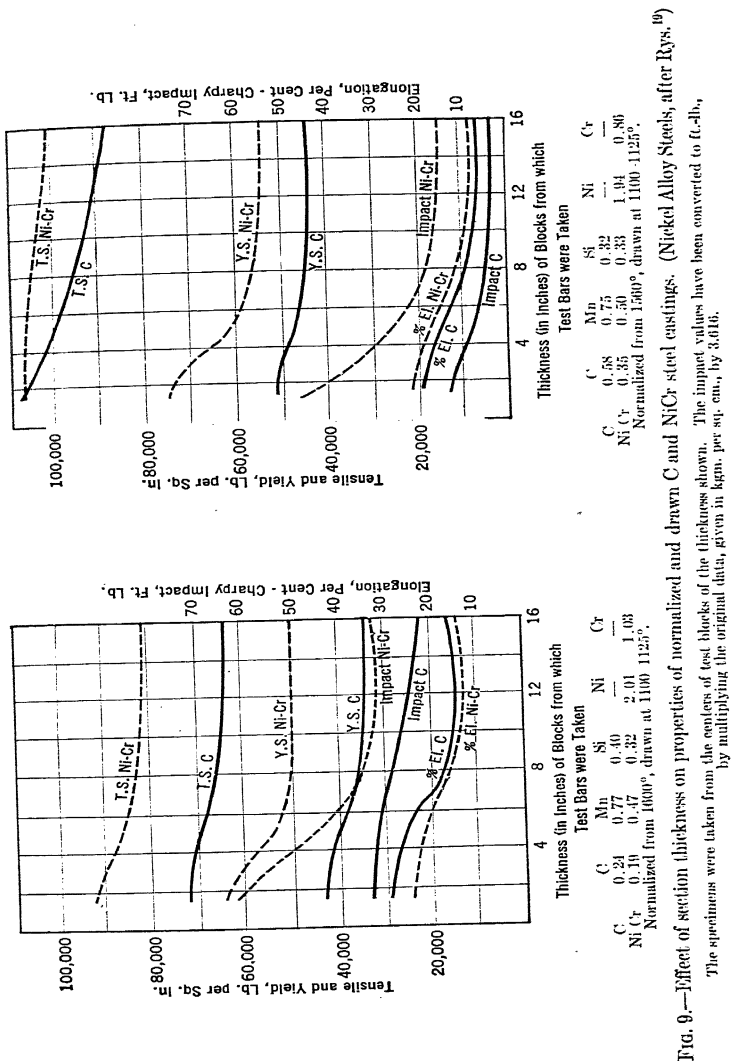
See general comment on property charts, p. 40, chap. 1.

**Mass Charts.** A picture of the depth hardening in Ni-Cr and Ni-Cr-Mo steels can be had from the mass charts for steels of this class given in the "S.A.E. Handbook," and in "Nickel Alloy Steels," a few of which are shown in Figs. 7, 8, and 9. Figure 7a for steels of similar Ni content shows how markedly Mo improves depth hardening, and Fig. 7b indicates that, even when the comparison is with a steel of higher Ni content, the Ni-Cr-Mo shows the greater depth hardening. The largest sections dealt with in Fig. 7 are 6-in. diam., with the specimens taken midway. The effect of depth-hardening elements is brought out in Figs. 8 and 9, in which 4- to 16-in. blocks of cast C and Ni-Cr steels are compared.

Fig. 8.—Effect of section thickness on properties of quenched and drawn C and Ni-Cr steel castings. (Nickel Alloy Steels, after Kys.<sup>19</sup>)

The specimens were taken from the centers of test blocks of the thicknesses shown. The impact values have been converted to ft.-lb., by multiplying the original data, given in kgm. per sq. cm., by 3.616.





**Huge Sections.** Data on still larger sections are available from the work of Maurer and Gummert.<sup>11</sup> Steel of 0.37 C, 0.38 Mn, 0.23 Si, 2.96 Ni, 1.15 Cr, 0.42% Mo was forged from 60-in. diam. ingots to cylinders of 40, 33, and 25.5-in. diam. and the cylinders hollow bored with holes  $\frac{1}{10}$  the outside diameter, leaving wall thicknesses of 1S, 1S, and 11.5 in., respectively. The hollow cylinders were oil quenched from 1600° and drawn at 1210°. In small sections this steel, so quenched and tempered, would show about 150,000 tensile, 130,000 yield, 20% elong., 60% R.A. In these large sections the properties ran about 110,000, 80,000, 10-15, 20-30. For a 0.55 C, 0.50 Mn, 1.50 Ni, 0.75 Cr, 0.30% Mo steel, Finkl, quoted by Gregg,<sup>2</sup> gives in a 17-in. diam. section, normalized and drawn instead of quenched and drawn, 120,500, 90,500, 20, 42 on specimens taken midway. Gregg also cites a steel of 0.30 C, 3.00 Ni, 1.50 Cr, 0.50 Mo, 0.20% V steel in 10- to 15-in. diam. sections, air quenched and tempered, as giving 115,000, 95,000, 1S, 52. That is, in very large sections nothing is gained by quenching over normalizing. Therefore, the steels are designed for normalizing. For forgings of the order of 40-in. diam. Mochel<sup>1</sup> cites the following, on which liquid quenching is not permitted, normalizing and drawing being required.

TABLE OF CHEMICAL COMPOSITION AND TENSION TEST REQUIREMENTS OF TYPICAL TURBINE ROTOR FORGING SPECIFICATIONS

CHEMICAL COMPOSITION							
TYPE	1	2	3	4	5	6	7
Carbon.....	0.48 max.	0.45 max.	0.43 max.	0.40 max.	0.43 max.	0.43 max.	0.43 max.
Manganese.....	0.40-0.80	0.40-0.80	0.40-0.80	0.75 max.	0.40-0.80	0.40 max.	0.90 max.
Phosphorus.....	0.04 max.	0.04 max.	0.05 max.	0.04 max.	0.04 max.	0.04 max.	0.05 max.
Sulphur.....	0.045 max.	0.045 max.	0.05 max.	0.04 max.	0.045 max.	0.05 max.	0.05 max.
Silicon.....	.....	.....	.....	0.30 max.	.....	.....	.....
Nickel.....	.....	2.75-3.75	2.50 min.	3.25 max.	2.50 min.	2.50 min.	2.50 min.
Chromium.....	.....	0.60 max.	0.30-0.60	1.25 max.	0.60 max.	0.60 max.	0.60 max.
Molybdenum.....	.....	.....	0.20 min.	0.30-0.50	0.30 min.	0.30 min.	0.25 min.
Vanadium.....	.....	.....	.....	0.25 max.	0.25 max.	.....	.....

TENSION TEST REQUIREMENTS

TYPE	1	2	3	4	5	6	7
Tensile strength, lb. per sq. in.	75,000	55,000	90,000	85,000	95,000	100,000	80,000
Yield strength, lb. per sq. in.	40,000	55,000	65,000	.....	.....	.....	.....
Proof stress, lb. per sq. in.	.....	.....	.....	60,000	65,000	.....	.....
Elongation in 2 in. ....	20.0	20.0	18.0	15.0	18.0	18.0	18.0
Red. of area, per cent.....	35.0	40.0	40.0	24.0	35.0	40.0	40.0

The tensile properties shown above are for tangential test specimens for Types 1-2-3-5-6 and for radial specimens for Type 4, all located at greater than half radius of the forging. Longitudinal specimens require higher percentages of elongation and reduction of area than those shown, and, when tangential or radial specimens are located at less than half radius, slightly lowered values of elongation and reduction of area are allowed.

As will be seen in the next chapter, Ni-free, Mn-V or Mn-Cr-V steels are used for large normalized forgings, with properties of much the same general order. Even though we may wish to secure very high strength at the center of large sections and to this end employ highly alloyed, deeply hardening steels, perhaps so highly alloyed as to be air hardening, this is not always sound engineering. Difficulties arise in machining the very strong steels, for the strongest ones cannot be softened as much as the machine shop might wish. Even more important in the case of complex and irregular sections is the fact that in producing martensite even by air quenching, dimensional changes occur and quenching stresses are introduced. An undetected quenching crack in a large object subjected to severe duty, as is normally the case with large parts in which very high strength is sought, may be extremely dangerous. Discretion is often the better part of valor in design of huge parts. It may be much wiser to avoid Ar'', the martensite transformation, and utilize Ar' the aim being to produce finely emulsified pearlite, approaching primary troostite. That is, normalizing is resorted to instead of quenching.

In the steels that are air hardening in small sections the boundary between normalizing and air quenching is hazy. True normalizing involves the direct change of austenite to pearlite or primary troostite; true air quenching involves the change of austenite to martensite.

**Normalizing Methods.** To avoid the latter but still to accelerate the cooling so that the cooling rate at the center of a large mass shall be rapid enough to give fine rather than coarse pearlite, without letting the outside cool so fast as to become martensitic, requires high initial temperature and avoidance of rapid cooling through the martensite range, that is, a type of normalizing approaching "austempering." By air cooling, followed, before the piece cools far enough to enter the martensite stage, by placing in a low-temperature furnace, or by opening the furnace doors and blowing in air at first, then closing the doors to get slow cooling at the lower temperature range, this may often be accomplished though the method has to be specially worked out for each particular case.

Even in pure normalizing, internal stresses are not entirely avoided and it is necessary to draw back to remove them. Large turbine shafts are often subjected to repeated normalizings and temperings, including several repeated tempering reheatings, one or more of which, at a temperature higher than the shaft will operate at and continuing for many hours, is given after machining to final dimensions, to ensure complete permanence of dimensions.

Mochel's <sup>2</sup> discussion of this heat treatment has been referred to in Chap. 11, Vol. I.

**Nickel-Chromium Steels.** Although Mo or V, or both, are frequently used to assist the Ni-Cr steels to greater depth hardening, the Ni-Cr-Mo and Ni-Cr-V steels are specialty products, generally made up as special heats, on special order. Relatively few heat-treated parts are so massive as to require the extreme degree of depth hardening. For an intermediate degree, the unmodified Ni-Cr steels are in sufficient use so that they are given numbers in the S.A.E. 3100, 3200, 3300, and 3400 series.

The heat treatment of these Ni-Cr and more complex steels reflects the fact that  $A_{c2}$  is lowered by the presence of Ni, so that much lower

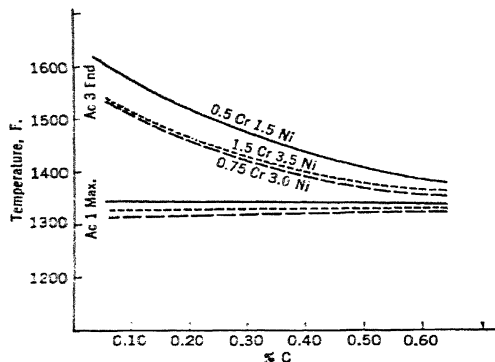


FIG. 10.—Critical points on heating of Ni-Cr steels.

The curves for 1 Cr 1.75 Ni are close to those for 0.5 Cr 1.5 Ni.

temperatures are required on heating than with a C steel. With the more highly alloyed steels of the group Ar' and Ar'' are lowered on cooling, and in small sections it is easy to cool fast enough to pass through Ar' and leave only Ar'', the martensite change. The steels, save in the very lowest C contents, are oil quenching rather than water quenching, and in the more highly alloyed ranges they are air hardening.

$A_{c2}$  and  $A_{c1}$  for straight Ni-Cr steels are shown in Fig. 10.

**Property Charts for Nickel-Chromium Steels.** Property charts for all the S.A.E. steels containing Ni, in all the various C, Ni, and Cr contents, and for both water and oil quenching when either treatment is applicable to the composition, are given in "Nickel Alloy Steels,"<sup>12</sup> which is obtainable from the International Nickel Company, New York. The many charts available in the literature will be reviewed and correlated in the forthcoming monograph on "Alloys of Iron and Nickel," being prepared by the Alloys of Iron Committee. It will

here to show in Figs. 11 and 12 charts for steels in the 0.30–

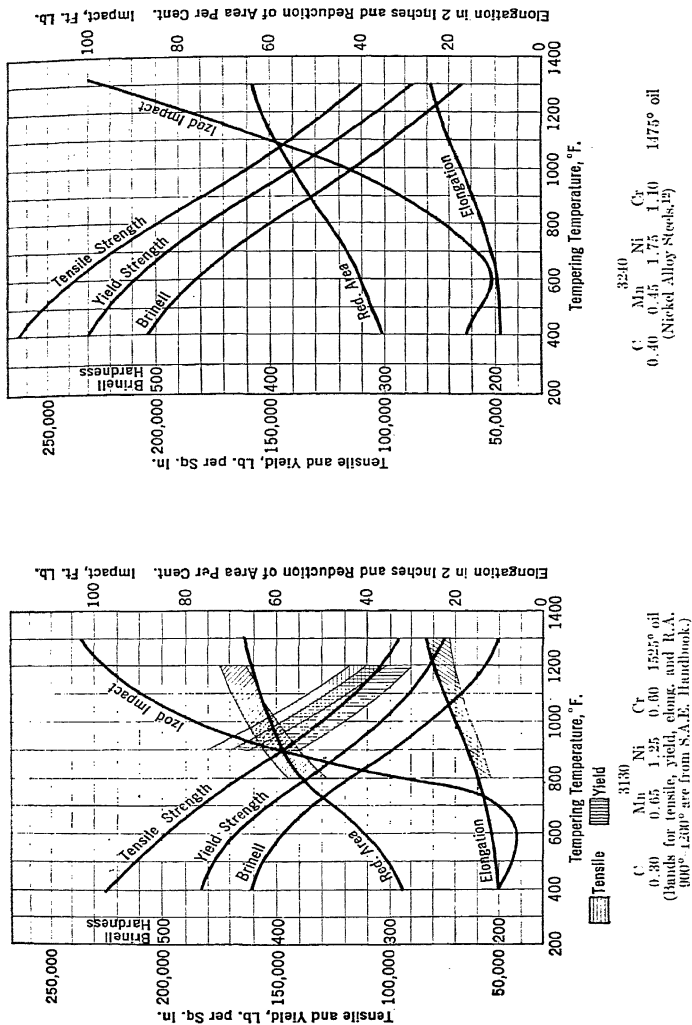


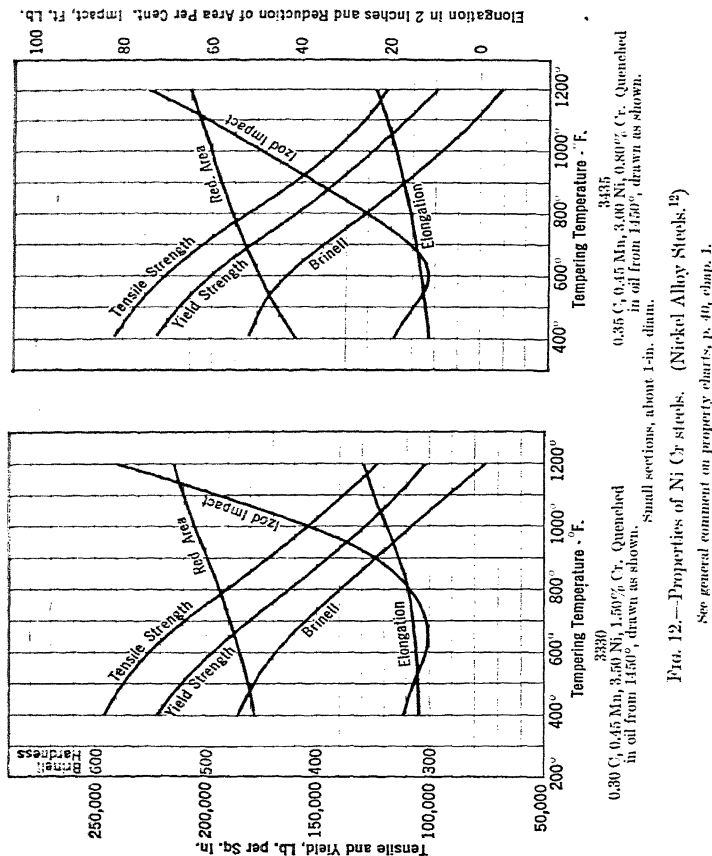
Fig. 11.—Properties of NiCr steels, in small sections, about 1 in., oil quenched and drawn as shown.

See general comment on property charts, p. 40, chap. I.



0.40% C range of the various S.A.E. groups, i.e., 3130, 3240, 3330, and 3435, oil quenched.

The bands for 3130 that were discussed in Chap. 1, Vol. I, have



been added to the chart for that steel. Data for 3140 were shown in Fig. 19, Chap. 1, Vol. I. The low-C carburizing steels will be discussed in Chap. 14. The usual heat treatment for Ni-Cr steels is referred to at the end of Chap. 10.

**Nickel-Molybdenum Steels.** Since Mo is a carbide-forming element, Ni steels with Mo, but without Cr, are somewhat analogous to Ni-Cr steels. The Ni-Mo steels are notable for good impact and reduction of area as well as high yield ratio. A steel of 0.30–0.35 C, 4.50 Ni, 0.60% Mo was used for light armor plate on tanks during the World War, for its combination of high yield strength and toughness, but most of these steels for other purposes use lower amounts of both Ni and Mo. The S.A.E. series includes:

	C	Mn	Ni	Mo
4615	0.10–0.20	0.40–0.70	1.65–2.00	0.20–0.30
4620	0.15–0.25	"	"	"
4640	0.35–0.45	0.50–0.80	"	"
4815	0.10–0.20	0.40–0.60	3.25–3.75	0.20–0.30
4820	0.15–0.25	"	"	"

Of these, the lower C members, primarily carburizing steels, will be discussed in Chap. 14. These steels are usually specified fine grained. They are especially useful in gears because of low distortion on quenching. The higher C 4640 used for oil hardening gears, with or without a cyanide or gas carburized skin, is favored for the same reason. The properties are shown in Fig. 13.

Davis<sup>13</sup> describes the use of 4640 at a lower draw than is shown in Fig. 13. This steel is valuable to the gear maker largely because of its very small distortion on quenching. In properties, for rear axle gears and the like, Davis considers cyanided 4640 equivalent to carburized 2512 or 4815. He specifies the usual S.A.E. analysis, save for restricting the C to 0.35–0.40%, with 6 to 8 grain size.

The steel is annealed for machinability to 195–215 Brinell at 1550–1750° and cooled at about 150° per hr. to 1200°. It is then cyanide cased to about 0.006 in. deep, oil quenched and drawn at 400°. The case is file hard, the core is about 53 Rockwell C, and runs 291,000–294,000 tensile, 181,500–185,500 yield, 10–12.5% elong., 30.5–36% R.A. Cyanide cased, V-notched Izod bars show only 1 to 3 ft.-lb. Bars heat treated but not cyanided show 5 to 6. Davis' comment that this steel shows "high strength and hardness without brittleness" is interesting in view of the insistence often laid on high impact.

Drummond's<sup>14</sup> data for 4640 vary a trifle from those of Fig. 14. He gives 4640, normalized from 1650 to 1700°, oil quenched from 1450 to 1500°, drawn as shown:

Draw	Tensile	Yield	Elong. %	R.A. %
800'	178,000	147,000	10	46
900'	158,000	130,000	11	48
1000'	142,000	115,000	13	51
1100'	128,000	104,000	16	55
1200'	119,000	94,000	19	58

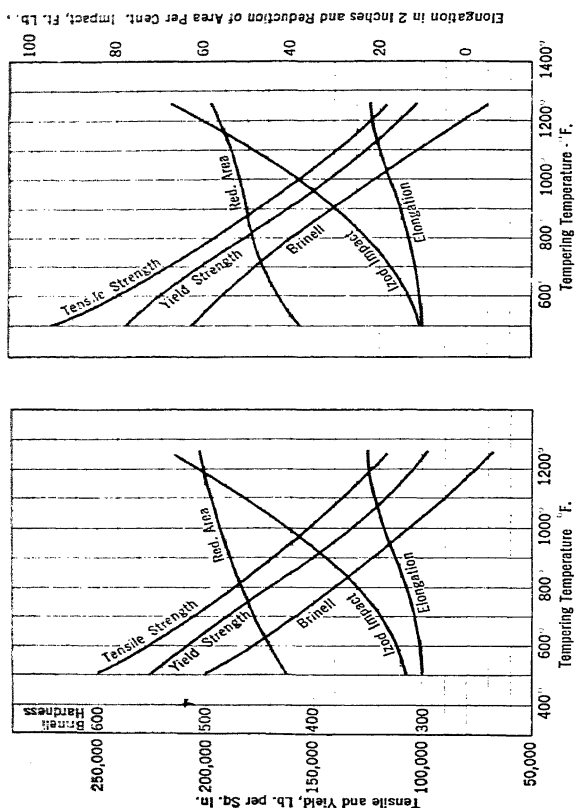
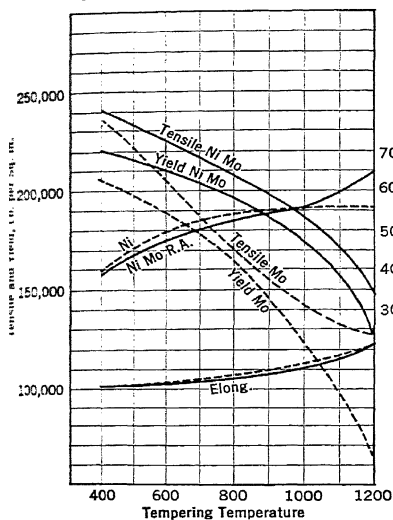


Fig. 13.—Properties of NiMo steels. (Nickel Alloy Steels,<sup>12</sup>)  
 0.40 C, 0.05 Mn, 1.80 Ni, 0.25% Mo. Quenched in oil from 1300°, drawn as shown. Small sections, about 1-in. diam.  
 0.50 C, 0.05 Mn, 1.80 Ni, 0.25% Mo. Quenched in oil from 1175°, drawn as shown. 4650

See general comment on property charts, p. 40, chap. 1.

The effect of Mo in the high-Ni steel for light armor plate is shown in Fig. 14.

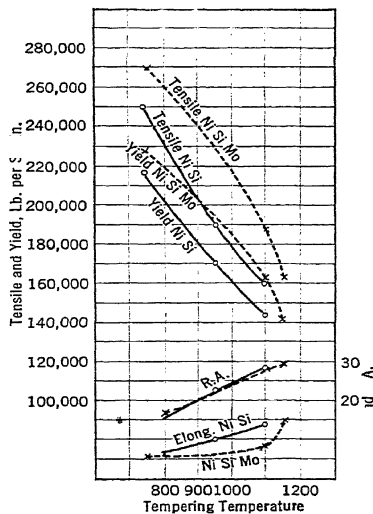
**Nickel-Silicon and Nickel-Silicon-Molybdenum Steels.** The ferrite-strengthening effect of Ni may be increased by Si. Properties of Ni-Si and Ni-Si-Mo steels, also used for light armor plate, are shown in Fig. 15. The steel utilizing Si, without Ni, as ferrite strengthener falls into the spring steel (Chap. 16) or the tool and die steel classes (Chap. 22).



1½ in. square bars, oil quenched from 1500°. (Sargent.<sup>10</sup>)  
Solid line: 0.33 C, 0.25 Mn, 4.50 Ni, 0.58% Mo.  
Dashed line: 0.33 C, 0.52 Mn, 8.00% Ni.

Fig. 14.—Effect of Mo in high Ni steel.

See general comment on property charts, p. 40, chap. 1.



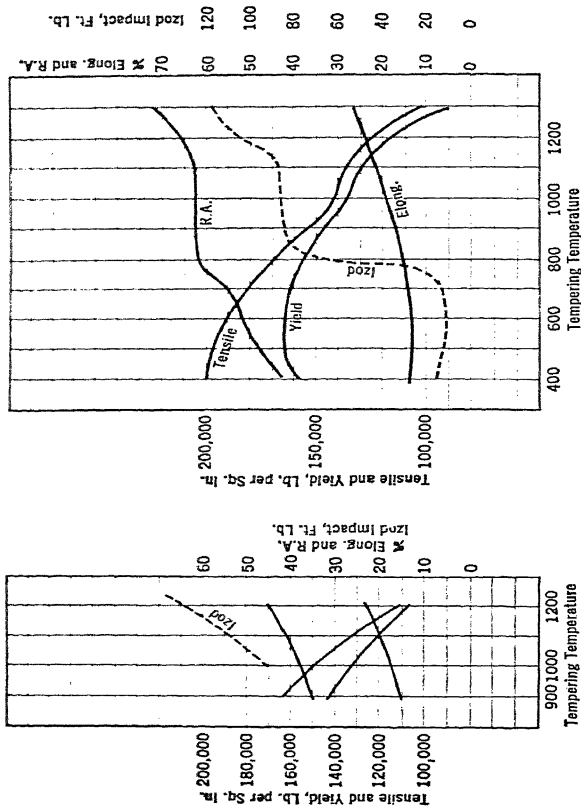
Treated and tested as flat bars, ¼ in. thick.  
Oil quenched from 1575° (Gillett and Mack.<sup>11</sup>)  
Solid line: 0.41 C, 0.70 Mn, 3.00 Ni, 2.47% Si.  
Dashed line: 0.37 C, 0.52 Mn, 2.95 Ni, 2.50 Si, 0.70% Mo.

Fig. 15.—NiSi and NiSiMo steels.

**Manganese-Chromium, Manganese-Chromium-Vanadium, and Manganese-Vanadium Steels.** Naturally, other ferrite strengtheners than Ni, and other carbide formers than Cr, can be combined to produce steels analogous to the Ni-Cr family. For example, Mn may be used as the ferrite strengthener instead of Ni.

Manganese-V steel has been discussed, for normalized castings, in Chap. 3. "Vanadium Irons and Steels" cites large forgings of 0.22–0.30 C, 0.15–0.30 Si, 1.45–1.75 Mn, 0.15% V, normalized 1575–1625° and drawn, 1000°–1200° as giving:

	Tensile	Yield	Elong. %	R.A. %	Izod
Center of 4-in. $\times$ 4-in. ....	96,500	78,500	29	63	55
Center of 6-in. diam. ....	97,500	71,500	26	57	49
Midway in 10-in. diam. ....	93,000	64,000	25	58	....
Midway in 13 $\frac{1}{2}$ -in. diam. ....	91,000	66,000	32	67	73
Midway in 14-in. diam. ....	91,500	64,000	25	57	....



(a) 0.27 C, 1.71 Mn, 0.25% Mo. Water-quenched from 1550°. Specimens from wall of tube 0.35 in. thick.

(b)

0.23 C, 1.68 Mn, 0.18% V. Normalized from 1050°, then water quenched from 1000°. Specimens 1-in. diam. (Vanadium Steels and Irons.)

Fig. 16. Properties of MnMo and MnV steels.

See general comment on property charts, p. 40, chap. 1.

A property chart for small sections of such a steel, water quenched and drawn, is shown in Fig. 16a.

**Manganese-Chromium-Vanadium.** A steel of this type specially designed for normalizing, known as "Normalloy," falls in this general category, since the Mn is high enough to serve as ferrite strengthener. It contains 1.00–1.60 Mn, 0.30–0.60 Cr, 0.06–0.12% V. Other grain-size control elements could be substituted for the V. The C is at 0.15–0.20% in carburizing grades, at 0.25–0.35% in grades for quenching and drawing, and at 0.35–0.50% for normalizing and drawing. Property charts for small sizes, water quenched and drawn, are given in "Vanadium Steels and Irons" for steel of 0.25, 0.30, and 0.40% C. That for 0.30% C is shown in Fig. 17.

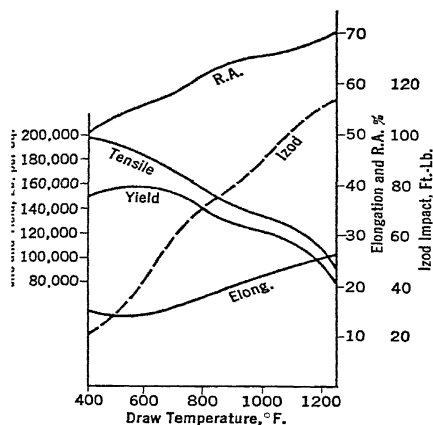


FIG. 17.—Properties of "normalloy" steel, 0.30 C, 1.20 Mn, 0.35 Cr, 0.10% V, normalized from 1625°, then water quenched from 1575°, drawn as shown. Treated in 1-in. diam. (Vanadium Steels and Irons).

See general comment on property charts, p. 40, chap. 1.

The normalized properties, determined on 1-in. rounds, are listed below:

C	Si	Mn	Cr	V	Tensile	Yield	Elong. %	R.A. %	Izod	Norm. from
0.25	0.28	1.20	0.44	0.03	81,000	55,000	32	64	78	1625°
0.30	0.25	1.20	0.35	0.10	106,000	68,000	26	62	60	1625°
0.41	0.27	1.19	0.34	0.11	112,000	78,000	27	62	.....	1600°

Data for similar steels of higher Cr content will be given in the next chapter.

**Manganese-Molybdenum.** Properties of a Mn-Mo steel are shown in Fig. 16a.

O'Neill<sup>11</sup> discusses a similar Mn-Mo steel, of 0.20–0.25 C, 1.50–1.75 Mn, 0.20–0.25 Si, 0.27–0.31% Mo, which, oil quenched from 1560° and drawn at 1110–1240°, gave 90,000–107,000 tensile, 71,500–85,000 yield, 23–31% elong., 57–70% R.A. The impact values varied from 37 to 85 Izod on coarse-grained and from 101 to 115 on fine-grained heats. O'Neill also described a weakly alloyed steel of 0.35 C, 0.95 Mn, 0.25 Si, 0.45 Ni, 0.05–0.10% Cr. Oil quenched from 1560° and drawn at 1150–1200°, this showed:

	Tensile	Elong. %	R.A. %	Izod
Coarse grained 0.0001% Al .....	70,000	27	64	12
Fine grained 0.01% Al .....	66,000	30.5	66	93

The advantage of grain-size control of the steels of this group is evident. When Cr or Mo is omitted and the Mn is kept under 1%, with V added, the Mn-V steel is generally termed "vanadium" steel. With Mn over 1%, and V added for grain-size control, the steel may be considered as somewhat analogous to the low-Ni, low-Cr series, since both ferrite strengthening and carbide forming tendencies are exhibited by the alloying elements. Data for 0.23 C, 1.68 Mn, 0.25 Si, 0.18% V steel, normalized from 1650° and water quenched from 1600°, are given in Fig. 16b. The ductility and impact values are noteworthy. Such steels have been discussed, as castings, in Chap. 3.

**Copper-Manganese.** Since Cu is a ferrite strengthener and Mn divides between ferrite and carbide, a Cu-Mn steel that might be considered for large sections may be mentioned in this chapter.

The usual minimum specifications for very large forgings are tensile 90,000, yield 60,000, elong. 25%, R.A. 50%.

These specifications should be cheaply met in a precipitation-hardened (normalized and reheated) Cu-Mn steel. One would expect a steel of 0.25 C, 1.25 Mn, 1.25% Cu, to behave as follows:

	Tensile	Yield	Elong. %	R.A. %
Normalized .....	80,000	45,000	30	65
Precipitation-hardened .....	100,000	65,000	25	50

Such a steel should have a price advantage over those commonly used for normalizing. Although some surface checking might be met in forging such a steel, large forgings generally have sufficient allowance for finish to insure removal of the checked zone. Precipitation hardening goes on uniformly through a large cross section, since the cooling rate necessary to hold Cu in solution, ready for the precipitation treatment, is so low that there would be little or no gradation from center to surface like the gradation from coarse pearlite to fine pearlite in large forgings of the regular normalizing steels. With suitable Al, Ti, or V addition for grain-size control, the Cu-Mn steels should be suitable for extremely large sections, even beyond the size that can be effectively strengthened in other alloys.

Harrison<sup>16</sup> made a comparison of 3½-in. diam. forgings, air cooled from 1560° and reheated 2 hr. at 930°. This is a precipitation-hardening temperature for the Cu steels.

## CENTER OF FORGING

No.	Composition, %					Tensile	Yield	Elong. % in 4 in.	R.A. %	Izod
	C	Si	Mn	Ni	Cu					
27, Cu	0.30	0.24	0.60	.....	1.06	95,000	72,000	29	51	19
28, Ni	0.26	0.23	0.62	1.40	.....	78,500	54,000	35	58	54
29, Cu-Ni	0.26	0.22	0.72	1.49	1.00	95,500	74,500	29	52	30
33, Mn-Cu	0.27	0.21	1.48	.....	1.05	107,000	83,000	28	51	33
34, Mn-Cu	0.29	0.21	1.90	.....	1.00	124,000	90,000	21	43	17

## OUTSIDE OF FORGING

No.	Tensile	Yield	Elong. % in 4 in.		Izod
27, Cu. . . .	97,500	75,000	27	51	17
28, Ni. . . . .	80,000	54,000	37	58	50
29, Cu-Ni.	101,000	83,000	30	53	31
33, Mn-Cu	109,000	86,500	28	51	28
34, Mn-Cu	123,000	96,500	25	51	21

When the 3½-in. diam. forgings were oil quenched from 1560° and drawn 2 hr. at 1110°, i.e., at a temperature at which over-aging of Cu steel is well advanced, the following results were obtained:



## 248 COMPLEX ALLOY STEELS BASED ON FERRITE FORMERS

## CENTER OF FORGING

No.	Tensile	Yield	Elong. % in 4 in.	R.A.	Izod
27, Cr.....	94,000	72,000	25	51	29
28, Ni.....	99,500	54,000	32	64	65
29, Cr-Ni.....	98,500	78,500	31	58	40
33, Mn-Cr.....	99,500	77,500	25	49	29
34, Mn-Cr.....	119,000	96,000	18	43	15

## OUTSIDE OF FORGING

No.	Tensile	Yield	in 4 in.	R.A. %	Izod
27, Cr.....	96,500	71,000	21	55	38
28, Ni.....	83,500	58,500	35	64	65
29, Cr-Ni.....	100,000	81,500	30	57	40
33, Mn-Cr.....	101,500	81,000	21	55	31
34, Mn-Cr.....	122,500	103,000	24	45	20

**General Behavior.** The ferrite-strengthened steels, in which the properties are augmented by carbide-forming elements, dealt with in this chapter, are especially suitable where strong depth hardening is required. The Ni-Cr steels are typical of this group. At medium-C contents, the more highly alloyed steels are air hardening in small sizes, and when quenching is resorted to, it is usually in oil.

Because of the air-hardening tendency, an ordinary anneal is likely to leave them too hard to machine rapidly, so hardening followed by tempering at above 1200° is often resorted to to put them in condition for machining.

Even air cooling is likely to leave hardening stresses in these steels. Tempering back with considerable softening, or, in "spring temper," a very long draw at a low draw temperature, is necessary to avoid retention of stress. The steels, especially those containing Mo or V, are resistant to tempering, and soften only slightly and very slowly at low draws. The retention of internal stress is so marked that, in a series of draws at increasing temperatures, the yield strength and the endurance limit often rise to a maximum before they start to fall, instead of falling steadily with increase in draw temperature.

In any of the steels of this group the grain size may be kept fine

and the impact thereby improved, by the addition of a small amount of V, or other suitable element, for grain-size control.

Although all the steels discussed in this chapter have useful properties in small sizes, their special utility is in large sizes where their depth-hardening properties may be utilized. The alloy content is adjusted so as to suffice for depth hardening in the size to be treated, i.e., the lowest alloy content that will serve will ordinarily be selected.

The critical cooling rate is materially affected by very small variations in C or by rather small variations in the carbide-forming elements. The content of Ni is not so critical, and varies between 1.50 and 3.50% in the majority of the steels of this group. The more highly alloyed steels are more commonly used as large forgings rather than as commercial bar stock.

Huge forgings will normally be made up from special electric furnace or moderately small acid open-hearth furnace melts so the composition can be ordered especially for the job without undue cost or hardship. The ingot size must be chosen with respect to the cross section of the forging and the amount of reduction to be put upon it, so that special rather than stock ingots are normally required. Since ingots of these types of steels would not be stocked normally, this field is one of special orders without too much standardization of composition to limit the user's choice.

Hence, although there are important tonnage steels in the group, it is of greatest interest from the point of view of special "tailor-made" steels for specific and exacting purposes, where the price of the steel is of secondary importance to securing the optimum combination of properties.

#### BIBLIOGRAPHY

1. MOCHEL, N. L.: Steel for Turbine Rotor Forgings and Their Heat Treatment, *Metals & Alloys*, v. 8, 1937, pp. 265-269.
2. GREGG, J. L.: "The Alloys of Iron and Molybdenum." New York, McGraw-Hill Book Co., 1932, 507 pages.
3. Vanadium Corporation of America: "Vanadium Steels and Irons." New York, Author, 1937, 189 pages.
4. STRAUSS, J.: Medium Carbon Pearlitic Manganese Steels, *Trans. Am. Soc. Steel Treat.*, v. 14, 1928, pp. 1-26.
5. GREAVES, R. H.: Properties of Some Low-Nickel Steels Containing Manganese, *J. Iron Steel Inst.*, v. 132, 1935, pp. 99-116.
6. FINKL, W. F.: Die Block. U. S. Patent 2,104,979. January 11, 1938.
7. JONES, J. A.: The Influence of Molybdenum on Medium Carbon Steels Containing Nickel and Chromium, Res. Dept., Woolwich, *R. D. Rep.* 67, 1926, 59 pages. Abst.: *Chem. Age*, Monthly Met. Sect., v. 16, 1927, pp. 17-19. Abst.: *Mech. Engng.*, v. 49, 1927, p. 686.

## 250 COMPLEX ALLOY STEELS BASED ON FERRITE FORMERS

8. BOEGEHOLD, A. L.: Plain Carbon Steel, *Metal Progress*, v. 31, February, 1937, pp. 147-152.
- BOEGEHOLD, A. L.: Advantages of Alloy Steels, *Ibid.*, v. 31, March, 1937, pp. 265-269.
- BOEGEHOLD, A. L.: Appraising a Steel for a Given Duty, *Ibid.*, April, 1937, pp. 403-406.
9. ABRAHAM, H. H.: The Influence of Vanadium on Nickel-Chromium and Nickel-Chromium-Molybdenum Steels, *J. Iron Steel Inst.*, v. 134, 1936, pp. 241-249.
10. KINZEL, A. B. and W. CRAFTS: "The Alloys of Iron and Chromium." New York, McGraw-Hill Book Co., Vol. 1, 1937, 535 pages.
11. MAURER, E. and H. GUMMERT: Gefüge und Festigkeitseigenschaften grosser Schmiedestücke (Structure and Mechanical Properties of Large Forgings), *Stahl u. Eisen*, v. 54, 1934, pp. 1281-1289, 1309-1320.
12. FRENCH, H. J., and J. W. SANDS (editors): "Nickel Alloy Steels." New York, International Nickel Co., Inc., 1934, 7 Sections.
13. DAVIS, E. F.: "4640" The Unknown, *Automotive Industries*, v. 75, July 18, 1936, pp. 90-93.
14. DRUMMOND, R. S.: Automotive Gear Design Demands Modern Methods, *Automotive Industries*, v. 78, 1938, pp. 448-450, 459.
15. O'NEILL, H.: Alloy and Fine-Grained Steels for Locomotive Coupling Rods, *J. Iron Steel Inst.*, v. 135, 1937, pp. 187-221.
16. HARRISON, R.: The Effect of Copper on Some Alloy Steels, *Iron Steel Inst.*, Advance copy No. 4, May, 1938, 38 pages.
17. SARGENT, G. W.: The Value of Molybdenum Alloy Steels, *Trans. Am. Soc. Steel Treat.*, v. 1, 1921, pp. 589-597.
18. GILLET, H. W. and E. L. MACK: "Molybdenum, Cerium, and Related Alloy Steels." New York, Chemical Catalog Co., 1925, 295 pages.
19. RYS, A.: Legierter Stahlguss in Theorie und Praxis (Alloy Steel Castings in Theory and Practice), *Stahl u. Eisen*, v. 50, 1930, pp. 423-438.

## CHAPTER 13

### COMPLEX ALLOY STEELS BASED ON CARBIDE FORMERS— CHROMIUM AND SIMILAR FAMILIES

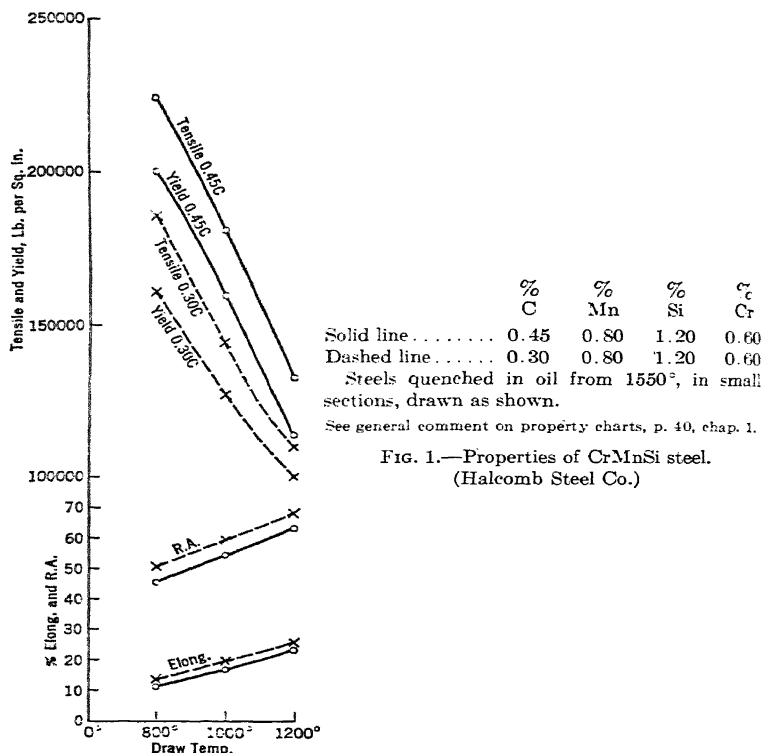
Chromium, as has been mentioned in Chap. 7, is a cheap alloying element which confers depth-hardening properties and enters cementite to produce a hard, wear-resistant carbide. It has some ferrite-strengthening power, but it is, on the whole, a hardening element rather than a toughening one. The constructional steels based on Cr seldom contain more than 1.50%, that is, as Fig. 7, Chap. 7 shows, they are pearlitic steels. The Cr is very commonly used along with larger quantities of other elements, notably Ni, sometimes Mn, as discussed in the last chapter.

In the constructional steels in which Cr predominates, other elements are added to secure certain advantages, usually to toughen without losing much hardness. Some special wear-resistant cast steels are based on Cr, so augmented; but the Cr cast steels with Cr predominating are relatively unimportant, since the necessary toughness can be had in other ways. The Cr steels are used primarily in wrought form.

**Chromium-Manganese.** Unless V, or some other grain-refining element is used, the Mn content of low-Cr steels, say 1% Cr, for quenching and tempering, or for normalizing, is seldom over 1%, for cracking is likely to occur on quenching, and too great air hardening on normalizing, if the sum goes over 2%. Up to around that limit, distortion is minimized and the combination is used in high-C, "non-deforming" tool steels. Ordinarily the Mn of Cr steels is held down so that only a slight added depth-hardening effect is ascribable to Mn, and the steels are generally thought of as plain Cr steels. Conversely, in the steels with over 1% Mn, the Cr is usually held under 0.50%, so as not to overdo the depth-hardening effect, in the absence of definitely ferrite-strengthening elements. "Normalloy," discussed in the latter part of the last chapter, also containing V, has the Cr so restricted. In Chap. 7, the beneficial effect of substituting 0.25% Cr for that much Mn has been commented upon. Chromium-Mn spring steels

are in use, but in these, also, Mn is only slightly increased over the normal content, despite the name.

When Cr is increased to 2% it appears better, on the score of impact and temper brittleness, to keep Mn below 0.50%, as Kinzel and Crafts<sup>1</sup> point out. Very careful balancing of Cr and Mn appears necessary.



**Chromium-Silicon and Chromium-Manganese-Silicon.** The effect of Mn+Cr is mitigated by raising Si. Steel with 1% each of the three elements has been studied, but the combination is not outstanding in properties in the quenched and tempered condition. In the as-rolled or normalized condition a steel with about 1.25 Mn, 0.75 Si, and 0.50%

Cr, with C usually rather low, called Cromansil, has found wide application. This has been discussed in Chap. 11.

Steel with lower Mn but with higher Si and Cr, a decidedly cheap composition, has been in commercial use. Properties of 0.30 and 0.45% C steel of this type, as given by the Halcomb Steel Co., are shown in Fig. 1. At the higher C and lower draw, this is a spring steel composition and has been used as such. Analogous steels with Si down to 0.75 and Cr up to 0.75% have been used for general constructional purposes.

Steels with around 0.35–0.50 C, 0.50–1.00 Mn, 1.75 Cr and 0.75–1.25% Si, reported by Greaves,<sup>2</sup> oil quenched from 1605° and drawn at 1200°, had higher strength and lower ductility than the steels of Fig. 1 at the same draw, but were decidedly temper brittle. At this level of Cr, the steel with 0.35 C, 0.60 Mn, 0.35% Si showed a better combination of yield strength and ductility at that draw than the more highly alloyed steels.

The Cr steels with the cheap alloying elements, Mn and Si, are therefore limited in their capabilities.

**Chromium-Copper Steel.** The substitution of some Cu for Ni in Ni-Cr steels for quenching and tempering was discussed in the last chapter. Low-C steels containing Cu and Cr have been mentioned in Chap. 11. Jones<sup>3</sup> normalized (1690°, 1580°) steel of 0.31 C, 0.07 Si, 0.46 Mn, 0.88 Cr, and 1.17% Cu in 1¾ by ¾-in bars, obtaining 91,000 tensile, 63,000 yield, 31% elong., 61% R. A., 55 Izod. Such a steel could be precipitation hardened. Buchholtz and Köster<sup>4</sup> tested a steel of 0.18 C, 0.01 Si, 0.75 Mn, 0.42 Cr, and 0.87% Cu using longitudinal specimens taken at the center of a 12-in. and a 4-in. diam. forging with the following results:

	Tensile	Yield	Elong. %	R.A. %	Charpy mkg. cm. <sup>2</sup>
Normalized, 12 in. ....	71,000	48,500	25	69	12
Normalized, precipitation hardened 12 in. (930°—4 hr.) ....	82,500	62,500	19	59	10
Normalized, 4 in. ....	72,500	52,500	26	69	12
Normalized and precipitation hard- ened, 4 in. ....	84,000	67,000	22	65	9

Higher Cu would give still greater increase in strength on precipitation hardening. Lequis, Buchholtz, and Schulz<sup>5</sup> water quenched a 1-in.

bar of 0.19 C, 0.01 Si, 0.75 Mn, 0.64 Cr, 1.02% Cu steel from 1560° and secured:

Draw	Tensile	Yield	mkg.		
900° 1 hr.	127,000	105,000	13	61	12
1110° 2 hr.	96,000	88,000	18	75	21

The 900° draw would develop precipitation hardening, the 1100° draw would over-age the steel, and the action of Cu with such treatment is merely that of a ferrite strengthener. Ordinarily the Cr-Cu steels are not used quenched and drawn.

**S.A.E. Steels.** The common additions to steels in which Cr predominates are the expensive elements V and Mo. The S.A.E. list includes the 6100 and 4100 series, as follows:

	C	Mn	Cr	V	Mo
6115	0.10-0.20	0.30-0.60	0.80-1.10	0.15-0.18	.....
6120	0.15-0.25	"	"	"	.....
6125	0.20-0.30	0.50-0.80	"	"	.....
to					
6150*	0.45-0.55	"	"	"	.....
6195	0.90-1.05	0.20-0.45	"	"	.....
4130†	0.25-0.35	0.50-0.80	0.50-0.80	.....	0.15-0.25
X4130	0.25-0.35	0.40-0.60	0.80-1.10	.....	"
4140	0.35-0.45	"	"	.....	"
4150	0.45-0.55	"	"	.....	"

\* By 0.05 steps. Chromium-V spring steels are in use with 0.45-0.75% C and Cr either as in 6115, or lower.

† 4120 with 0.15-0.25% C, although not on the S.A.E. list, is in use and is commonly referred to as though it were listed.

Of these, 6115, 6120, and 4120 are primarily carburizing steels; 6195 is primarily a ball-bearing steel. The 0.25-0.50% C grades are used for general constructional purposes.

**Chromium-Vanadium Steels.** The introduction of V into Cr steels is primarily for grain-size control, which might similarly be obtained by use of Al, Ti, or Zr, but the amount specified in the 6100 series is somewhat above the V content of 0.08-0.12%, usually considered to suffice for grain control. A small alloying effect of the extra 0.05-0.10% V seems to be exerted, the yield strength being raised more than would be expected from mere grain-size control. The most important

effect is in improving ductility and impact through keeping the grain fine, at normal temperatures of heating. The greater latitude in normalizing and quenching temperatures over that allowable in plain Cr steel is a major reason for the popularity of the 6100 series. The Cr-V steels have reasonably good properties as rolled or normalized, a steel of 0.25 C, 0.20 Si, 0.75 Mn, 1.00 Cr, 0.20% V in as-rolled 1.5-in.

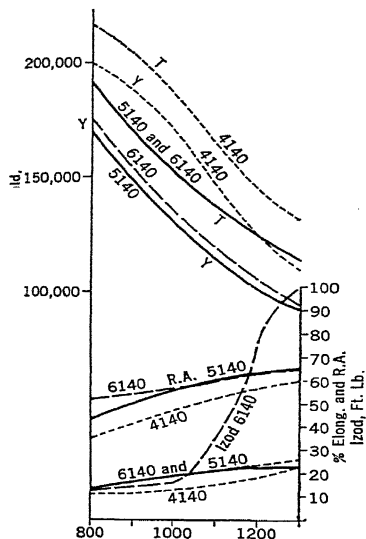


FIG. 2.—Properties of S.A.E. 6140, 5140 and 4140.

All have 0.40 C, 0.75 Mn, 0.20 Si, 1.00% Cr. 4140 also has 0.20% Mo, 6140 also has 0.18% V. 6140–1800° oil, 1-in. round. 5140–1550° oil, 1-in. round. 4140–1575° oil, 1-in. round. (Bethlehem Alloy Steels.) (The heavy dashed curves for Izod impact on 6140 from Vanadium Steels and Irons.<sup>6</sup>)

See general comment on property charts, p. 40, chap. 1.

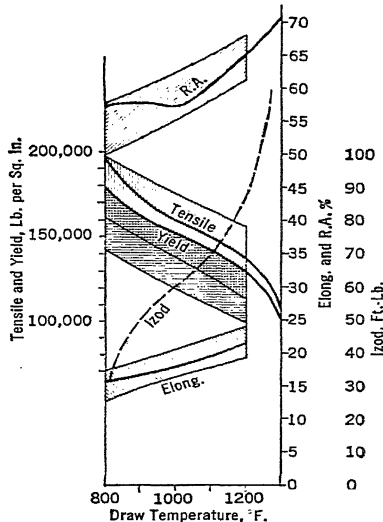


FIG. 3.—Properties of S.A.E. 6130 in 1-in. rounds.

Bands. 0.30 C, 0.75 Mn, 0.20 Si, 1.00 Cr, 0.18% V, quenched from 1800° in water, drawn as shown. (S.A.E. Jour. Jan. 1928, p. 55.) Lines. 0.33 C, 0.61 Mn, 0.18 Si, 1.03 Cr, 0.18% V. Normalized from 1625°, then quenched from 1575° in water, drawn as shown. (Vanadium Steels and Irons.<sup>6</sup>)

plate showing 104,000 tensile, 70,500 yield, 44% elong., 52% R.A., and a similar steel, normalized at 1650° in 1-in. diam., giving 85,000 tensile, 60,000 yield, 32% elong., 67% R.A., 102 Izod. One of 0.33 C, 0.18 Si, 1.03 Cr, 0.18% V, normalized in 1-in. round from 1625°, showed 107,000 tensile 85,000 yield, 31% elong., 67% R.A., 84 Izod; and one of 0.50% C, otherwise of the same composition, from 1600°, 127,000 tensile, 89,000 yield, 27% elong., 61% R.A., 26 Izod.



The full possibilities of the Cr-V steels are, of course, only realized by quenching and tempering, when the best combination of yield strength and ductility is sought. The secondary hardening effect due to V, shown in Fig. 4, Chap. 12, though not very marked with V under 0.20%, nevertheless does tend to offset the softening on tempering

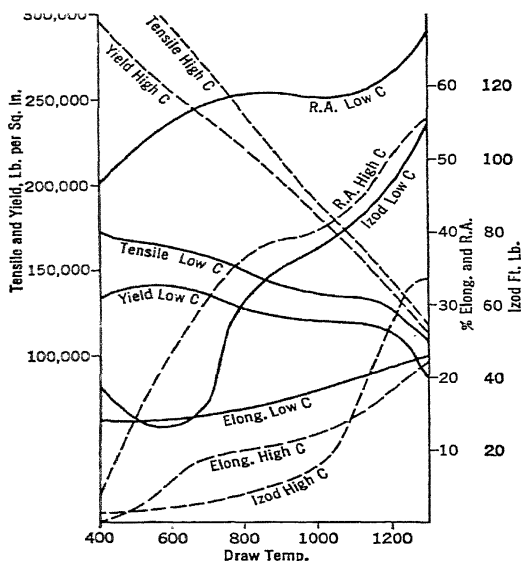


FIG. 4.—Properties of CrV steels of high- and low-C content.

	C	Mn	Cr	V	Normalized from	Quenched from	In
Solid lines	0.23	0.75	0.96	0.17	1650°	1625°	water
Dashed lines	0.63	0.72	0.98	0.17	1600°	1550°	oil

Drawn as shown. Bars 1-in. diam. or smaller. (Vanadium Steels and Irons.<sup>6</sup>)

See general comment on property charts, p. 40, chap. 1.

in the 900–1100° range, so that the properties are not as much affected by variation in draw temperature as in some other steels. That is, temperature insensitivity is met both in heating for quenching and, to a lesser degree, on tempering. Figure 2 shows how the V addition enhances the yield throughout and the R.A. at low draws, compared to a plain Cr steel of 0.40% C. Figure 3 shows the bands discussed in Chap. 1, Vol. I, obtained by the S.A.E. in its cooperative study of water-quenched 6130. Figure 4 shows data for water-quenched material of lower C content, and oil-quenched material of higher C content.

The 6130 is alternatively handled by oil or water quenching. The elong., R.A., and Izod of oil-quenched 6130 are but slightly below those shown in Fig. 4 for water-quenched 6125, whereas, except at the 1300° draw, the tensile and yield run 10,000 to 20,000 higher. Data on a steel of 0.26 C, 0.06 Si, 0.48 Mn, 0.92 Cr, 0.20% V quenched from different temperatures, in small sections, follow, together with, data from Burton and Russell<sup>20</sup> for one (B) of 0.26 C, 0.19 Si, 0.55 Mn, 0.66 Cr, 0.56 Mo, 0.30% V.

Quenched from, Deg.	Draw Temp. Deg.	Tensile	Yield	Elong. %	R.A. %
Water					
1560	840	202,000	173,000	12.5	54.5
1650	840	204,500	176,000	12.5	54.5
1560	930	177,000	165,000	14	59
1650	930	183,000	166,500	12.5	56.5
1560	1020	167,000	149,000	14	59
1650	1020	167,500	151,000	12	53.5
1560	1110	156,500	146,500	17	61
1650	1110	161,000	149,500	16	60.5
1560	1155	153,000	136,500	27	60
1650	1155	156,000	133,000	18	62.5
Oil					
1560	750	174,000	150,000	13	57
1650	750	183,500	155,000	13	51
1560	840	171,000	147,000	15	61
1650	840	177,500	151,500	14	53
1560	930	156,500	138,500	16.5	60
1650	930	162,500	146,000	15	57
1560	1020	137,500	113,000	21	64.5
1650	1020	141,500	123,000	18	63.5
1560	1110	133,000	108,000	17.5	65.5
1650	1110	137,000	112,000	20	61
1560	1155	131,000	100,000	28	67
1650	1155	133,000	99,000	30	70
As-rolled		132,000	110,000	19	51.5
Annealed at 475°		83,500	61,000	35	66.5
(B) Water					
1650	750*	192,000	139,000†	14.5	52
1650	890	179,000	139,000	17.5	54.5
1650	1065	183,000	141,000	18	58
1650	1155	166,000	148,000	19.5	59
1650	1245	125,000	110,000	24.5	67

\* Drawn 6 hr.

† Proportional limit.

## 258 COMPLEX ALLOY STEELS BASED ON CARBIDE FORMERS

Steel of 0.50 C, 0.06 Si, 0.92 Mn, 1.02 Cr, 0.20% V, oil quenched and drawn as shown below, gave the following, in small sections.

Quench, Deg.	Draw Deg.	Tensile	Yield	Elong. %	R.A. %
1520	750	260,500	240,000	8	24
1560		262,000	240,000	7	22
1650		264,500	239,500	6.5	17
1520	840	238,000	213,500	10	35.5
1560		234,000	216,000	9	28.5
		237,500	221,000	10	29.5
1650		227,000	217,500	10	35.5
1520	930	215,500	193,000	12	41.5
		214,500	201,500	11.5	36
1600		214,000	203,500	11.5	43
1650		210,000	192,000	12.5	42.5
1520	1020	183,500	177,000	14.5	47.5
1560		197,000	187,000	12.5	45
1600		205,500	187,000	12	45
1650		201,000	186,000	13	45.5
	1110	186,000	161,000	13.5	45.5
1650		187,000	170,000	15.5	45
As-rolled		153,500	124,500	12.5	37
Annealed 1475°		103,500	63,500	25.5	61.5

On the whole, the Cr-V steels are either improved, by, or at least not sensitive to, high quenching temperatures.

**Manganese-Chromium-Vanadium.** If the Cr is held down and V added for grain-size control, higher Mn might be used. Chandler and Dawe<sup>10</sup> report for water-quenched steels of about 0.32% C and 0.11% V, drawn at 900°, as follows:

Mn	Cr	Tensile	Yield	Elong. %	R.A. %	Izod
1.09	0.94	173,500	166,500	16	56.5	46
1.09	0.65	164,500	153,000	15.5	53.5	50
1.66	0.65	166,500	158,000	17	62	70
1.86	0.38	167,000	162,000	16	57.5	57

**Larger Sections.** A steel of 0.23 C, 0.10 Si, 0.58 Mn, 0.82 Cr, 0.17% V in 2.5 to 2½ rounds, oil quenched from 1650°, drawn at 1050°, gave from 121,000 to 126,500 tensile, 104,500 to 111,500 yield, 17 to 20% elong., 53 to 51% R.A. One of 0.29 C, 0.06 Si, 0.28 Mn, 1.00 Cr, 0.20% V in 6-in. rounds, water quenched from 1690°, drawn at 1155°, gave 115,000 tensile, 90,000 yield, 21% elong., 55% R.A., while the average on nearly 300 10-in. axles of 0.35 C, 0.50 Mn, 0.90 Cr, 0.22% V was 109,000 tensile, 81,500 yield, 21.5% elong., 58.5% R.A.

Data from "Vanadium Steels and Irons" <sup>7</sup> on very large sections of Cr-V steels are shown in Fig. 5.

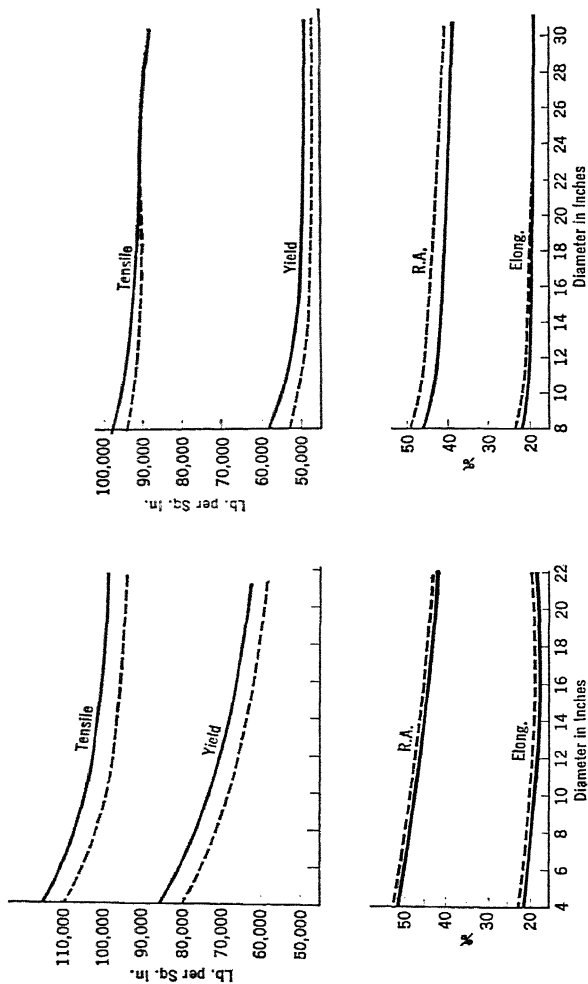
**Chromium-Molybdenum.** As Fig. 2 shows, the addition of Mo materially increases the strength of the Cr steels. This is also shown by the following from Houdremont.<sup>8</sup>

	% C	% Mn	% Cr	% Mo
A.....	0.35	0.70	1.00	....
B.....	0.33	0.70	1.10	0.20
C.....	0.30	0.50	1.40	0.30

For specimens oil quenched from 1545°, A-B drawn at 1125° (C at 1250°) taken at the center of bars about 0.75-in., 2.5-in. and 4-in. diam., he finds:

Diam. Inches		Tensile	Yield	Elong. %	R.A. %
¾	A Cr.....	117,000	105,000	22	66
	B Cr-Mo.....	134,000	114,000	22	66
	C Cr-Mo (high draw)...	109,500	85,000	25	71
2½	A Cr.....	109,000	74,000	24	62
	B Cr-Mo.....	115,000	91,000	22	63
	C Cr-Mo (high draw)...	104,000	79,000	25	73
4	A Cr.....	100,000	65,000	24	69
	B Cr-Mo.....	114,500	87,000	20	64
	C Cr-Mo (high draw)...	97,000	56,000	26	62

Camp and Francis<sup>9</sup> compared the properties of steels oil quenched from 1550° and drawn back to 125,000–128,000 tensile, as follows:



Oil quenched, drawn at 1200°. Dashed lines: 0.35 C, 0.40 Cr, 0.15 Mn, 0.90 Ni, 0.30 Si, 0.60 V. Solid lines: same except 0.40 0.45% C. Normalized, drawn at 1200°. (Vanadium Steels and Irons.)

Draw	C %	Mn %	Cr %	% Ni	% Mo	Tensile	Yield	Elong. %	R.A. %	Izod
1050	0.62	0.45	.....	.....	.....	126,000	84,500	18	43.5	5
1130	0.49	0.53	0.60	.....	.....	125,000	107,000	18	56.5	66.5
1000	0.40	0.65	.....	3.60	.....	128,000	112,000	19	51.5	54.5
1110	0.43	0.57	0.46	1.60	.....	128,000	111,000	20	60.5	54
1140	0.32	0.72	0.80	.....	0.27	125,800	112,000	21	68	90

The CrMo steel is outstanding in toughness.

**High Molybdenum-Chromium Steels.** Jones<sup>10</sup> reported on Cr steels with very high Mo, as follows:

C	Si	Mn	Cr	Mo	Quench	Draw	Tensile	Yield	Elong. %	R.A. %	Izod
0.25	0.07	0.32	1.10	1.38	1650° oil	1110	155,000	138,000	19	57	15
						1200	126,500	111,500	21	62	54
						1240	118,500	102,500	22	63	60
						1290	111,500	95,000	25	66	66
0.33	0.07	0.16	1.07	1.04	1650° oil	1020	181,000	158,500	16	52	18
						1110	159,500	136,500	18	55	23
						1200	141,000	124,000	20	62	49
						1240	133,000	116,500	21	64	59
						1290	118,000	105,500	25	66	70

Other data<sup>11</sup> for steels of high Mo, treated in 0.75-in. diam. after annealing at 1500°, oil quenching and drawing as shown, are:

C	Si	Mn	Cr	Mo	Quench	Draw	Tensile	Yield	Elong. %	R.A. %	Izod
0.25	0.19	0.48	0.95	0.73	1700° oil	800	187,750	150,000	15	56.5	33
						980	170,000	147,500	19	60	59
						1110	150,000	131,000	20	66	
						1165	119,500	100,000	22	71.5	95
0.40	0.40	0.63	0.95	0.68	1650° oil	800	219,000	170,000	12	43.5	19
						975	203,500	165,000	13	46	33
						1155	174,000	145,000	15	52	43
0.41	0.28	0.63	0.89	0.36	1650° oil	800	201,500	170,000	12	49	22
						975	178,000	155,000	14.5	52	33
						1155	140,000	112,000	19	62	71

Data for 0.25 C steel with 0.85 Cr and 0.35% Mo are shown in Fig. 6.

Increased strength and added resistance to tempering are conferred by Mo at these levels, but most CrMo steels now contain but about 0.20% Mo unless they are to be used at high temperature. Some resistance to grain growth and somewhat greater depth hardening than

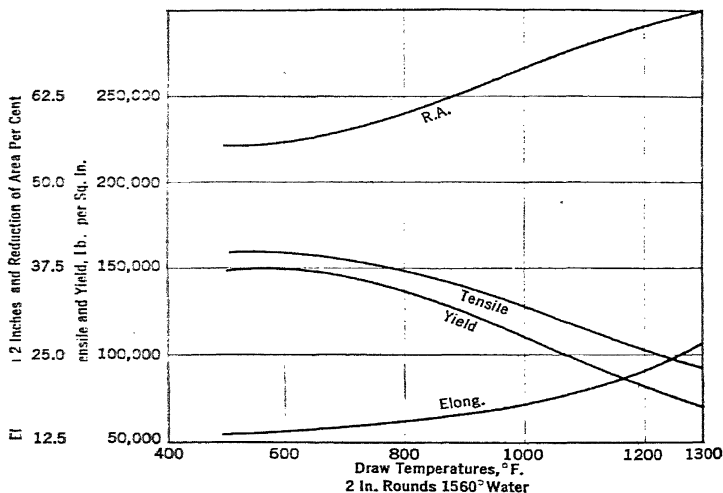
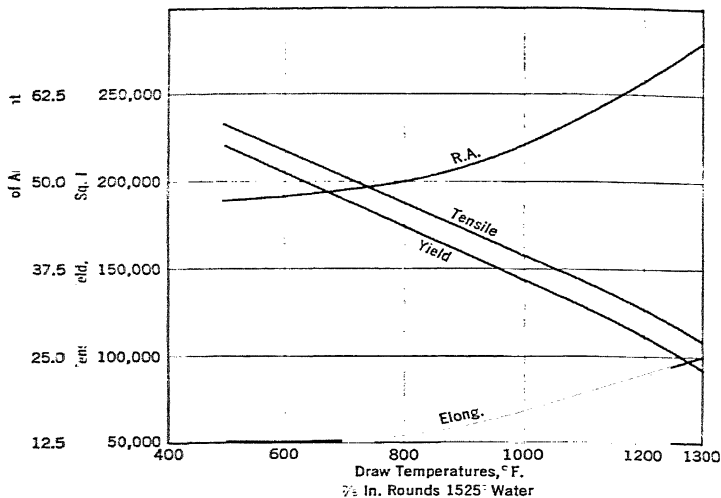


FIG. 6.—Properties of CrMo steel, in two sizes.

previously normalized from 1575°, water quenched and drawn as shown. 0.25 C, 0.10 Si, 0.75 Mn, 0.85 Cr, 0.35% Mo.

See general comment on property charts, p. 40, chap. 1.

in comparable plain Cr steels, with higher ductility at a given strength, are conferred by Mo at this level.

**Chromium Steels with Normal Molybdenum.** One steel in the S.A.E. series, 4130, carries less Cr, i.e., 0.50–0.80% instead of the 0.80–1.10% of the steels just cited. This is a water-hardening steel. Its properties, water and oil quenched, are shown in Fig. 7.

An important steel, X4130, has the usual 0.80–1.10% Cr, with about 0.20% Mo. The X4130 composition has found large use in

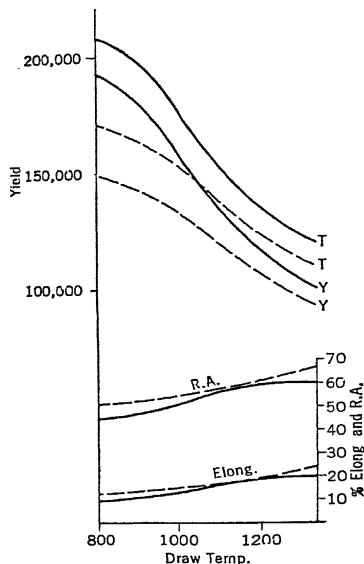


FIG. 7.—Properties of S.A.E. 4130.  
0.30 C, 0.20 Si, 0.65 Mn, 0.65 Cr, 0.20% Mo.  
Water quenched from 1550°, solid lines.  
Oil quenched from 1600°, dashed lines.  
(Bethlehem Alloy Steels.)

See general comment on property charts, p. 40, chap. I. In Republic Alloy Steels<sup>22</sup> quite variant results are given, the properties for 4130 water quenched being not far from those here shown for oil quenching. This example emphasizes the necessity for the comment on p. 40.

welded aircraft construction, wherein thin plates, sheets, and thin-walled tubing are employed.

In most welding, air-hardening steels are shunned since brittleness is feared. However, the approach to air hardening in this steel is not sufficient to cause embrittlement under suitable welding conditions, and normalizing, even from high temperatures, i.e., from a welding heat, produces properties not widely different from those after a quench and a 1200° draw, on thin stock. This has made it possible to assemble a very complex structure for aircraft use, by arc or oxyacetylene welding of tubing, and to rely on the properties of the air-cooled



welds, without necessity for heat treatment of the assembly. The rate of cooling from the welding heat is critical. The chilling effect of the mass of metal surrounding the tiny volume of an electric resistance spot weld is likely to be too great, so that spot welding of X4130 is avoided in order to prevent brittle joints. The welding behavior of some of the Cr-Mo steels has been discussed by several investigators.<sup>12-14, 14</sup> Tubing for aircraft use is of course made with the greatest care and rigorously inspected for uniformity of properties. In annealing between tube-drawing passes, controlled atmosphere furnaces are used to avoid decarburization.

The minimum properties of normalized X4130, tested as full tubes of different sizes, are listed by the Summerill Tubing Company as:

Wall Thickness	Tensile	Yield	Elong. in 2-in. % (Full Tube)
Up to 0.035 in.	95,000	75,000	10
0.036 to 0.186 in.	95,000	75,000	12
Over 0.186	90,000	70,000	15

(Similar values, or higher ones, are also reported by Whittemore and Bruggeman<sup>22</sup> for aircraft tubing.)

Such tubing is also used oil quenched and drawn, the following being minimum for tubes of  $\frac{1}{8}$  in. or over wall thickness, with the elongation increasing beyond the figures given, at greater wall thicknesses.

Draw	Tensile	Yield	Elong. in 2-in.
600°	200,000	150,000	5
800°	175,000	140,000	7
900°	150,000	130,000	9
1100°	125,000	110,000	11

These figures agree well with those determined by Sisco and Warner<sup>16</sup> for water-quenched and drawn X4130 in strip 0.063 in. thick. Johnson<sup>17</sup> reports the following for X4130 in  $\frac{1}{4}$ -in. plate:

Water Quenched from 1625°		Tensile	Yield	Elong. 2-in. %
Drawn	650°	222,000	160,000	9
"	900°	175,000	160,000	12
"	1100°	150,000	141,000	14.5

## Oil Quenched from 1625

Drawn	650°	229,000	204,000	9
	900°	178,000	165,000	12.5
	1100°	149,000	141,000	14.5

The similarity of properties on water and oil quenching, at the higher draws, is noteworthy.

The properties of 4140, which has the same Cr and Mo as X4130, have been shown in Fig. 1. For spring service, an alloy of 0.45 C, 0.75 Mn, 1.00 Cr, 0.30% Mo can be given the following properties, by quenching and drawing back to spring temper: 200,000–230,000 tensile, 180,000–210,000 yield, 12–15% elong., 35–45% R.A.

Houdremont's data, previously cited, showed improved depth hardening in 4-in. sections of 0.35 C, 1.00 Cr steel through addition of 0.20% Mo. For a 0.26 C, 0.57 Mn, 0.80 Cr, 0.36% Mo steel, water quenched and drawn, Dawe<sup>18</sup> found that the values for  $\frac{7}{8}$ -, 1-, and 1.5-in. diam. bars were quite uniform, but that in a 2-in. square bar the strength was lower.

	Draw	Tensile	Yield	Elong. %	R.A. %
$\frac{7}{8}$ -1.5 in.	1000°	157,500–162,000	147,000–154,000	19.5–21	56–57
2 × 2 in.		125,000	108,500	27.5	62
$\frac{7}{8}$ -1.5 in.	1100°	129,000–137,000	112,500–122,500	22.5–24.5	60.5–64.5
2 × 2		117,000	100,500	25.5	66.5
$\frac{7}{8}$ -1.5 in. }	1200°	112,500–119,000	93,000–101,000	26.5–27	66.5–69
2 × 2 in. }		100,500	79,000	29	70

The depth hardening in the commercial Cr-Mo steels, although marked, is not as great as in the commercial Ni-Cr and Ni-Cr-Mo steels. Compare Fig. 10, Chap. 10. The Mo-containing steels are resistant to tempering, but, when a temperature is reached at which they do start to soften, the change with increasing temperature is quite marked. (See Fig. 14, Chap. 10.)

Since 0.20% Mo costs about the same as 0.50% Ni, or 0.06% V, the Cr-Mo combination is the cheapest of the common ones in which Cr is helped out by another alloying element.

**Chromium-Molybdenum-Vanadium.** Greater depth hardening can be obtained by increasing C, Cr, and Mo and adding V. With such a steel the complete omission of Ni or other ferrite-strengthening elements is suggested by Houdremont as feasible for fairly large sections. Figure 8 for a steel of 0.45 C, 1.50 Cr, 0.50 Mo, 0.35% V in 2.5-in.

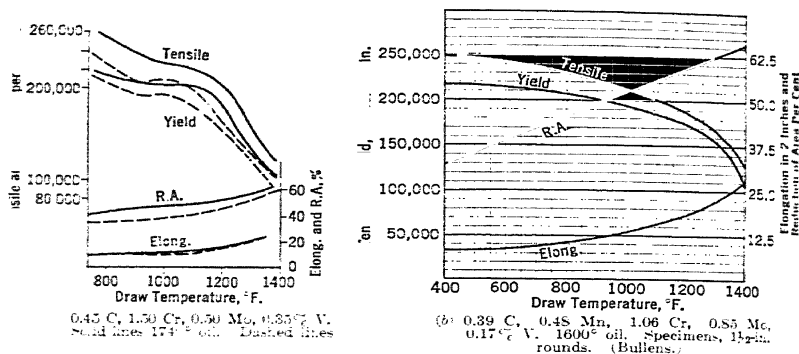


Fig. 8.—Properties of CrMoV steels.

(See general information on property charts, p. 40, chap. 1.

rounds shows that to reach the 22% elong., 60% R.A. ductility extremely high draw temperature is required. At lower draws, the yield strength is high but the ductility is none too good.

Data on an analogous Cr-Mo-V steel of lower C, i.e., 0.39 C, 0.60 Mn, 1.40 Cr, 0.35 Mo, 0.21% V, normalized from 1650° and either drawn directly after normalizing or oil quenched from 1650° and then drawn, are recorded in "Vanadium Steels and Irons," from tests on large sections, the figures cited below being for the center of 6-in. sections. All heating periods were 3 hr. at heat. Drawing was for 6 hr. at 900° and for 8 hr. in the other cases.

			Tensile	Yield	Elong. %	R.A. %	Izod
Norm.,	drawn	1100°	144,000	117,000	17	50	13.5
Quenched,	"	900°	164,500	137,500	15.5	52	15
"	"	1100°	164,000	137,500	17	56	16
"	"	1200°	138,000	121,000	19	59.5	44.5

The properties at the edge of the 6-in. diam. bar were also determined. They varied very little from those at the center. Figure 8b shows the properties of steel of similar C, but of a lower Cr and V and higher Mo level.

Other steels higher in Mo<sup>14</sup> gave the following for small sections:

C %	Mn %	Cr %	Mo %	V %	Oil Quench	Draw	Tensile	Yield	Elong. %	R.A. %
0.28	0.70	1.19	0.74	0.20	1575°	903°	204,000	190,000	13	48
					"	1100°	188,500	179,500	14	51
					"	1290°	125,000	120,000	20.5	60
					1600°	1200°	206,500	202,000	14	50
0.41	0.62	1.07	1.01	0.20	1690°	800°	253,000	220,000	10	37
					"	1000°	225,000	210,000	13	42
					"	1200°	208,000	195,000	14	47
					"	1400°	147,000	975,000	15.5	40

That the Cr-Mo-V combination may be considered for large normalized sections is shown in the following comparison, assembled from data in "Vanadium Steels and Irons" (except the last one, which is from Houdremont.<sup>8</sup>) For more highly alloyed steels, the normalizing and drawing treatments, especially the latter, were several times repeated. Entirely comparable data, from large objects of the same size and shape and from specimens taken at similar points, are not available, and allowance has to be made for the variation in section when such results are studied:

NORM. AND DRAWN						Location of Specimen	Diam. in.	Tensile	Yield	Elong. %	R.A. %	Izod
C %	Mn %	V %	Cr %	Mo %	Ni %							
0.30	0.80	0.15	.....	.....	.....	midway	7	97,000	61,000	25	53.5	.....
0.25	1.60	0.15	.....	.....	.....	center	6	97,500	71,500	26	57	49
0.45	0.80	0.15	1.00	.....	.....	midway	10	98,500	60,000	23	45	.....
0.39	0.60	0.21	1.40	0.35	.....	center	6	144,000	117,500	17	50	13.5
0.24	1.53	0.17	.....	0.21	.....	midway	6	97,000	77,000	25.5	67.5	53
0.33	0.83	0.11	0.74	0.41	1.85	center	19	120,000	90,000	17	37.5	.....
0.28	0.90	0.15	.....	0.35	2.92	"	13	107,000	81,500	24	60	.....
0.39	0.65	0.12	.....	0.34	2.87	"	12	115,000	94,000	23	57	.....
0.28	0.36	.....	1.26	0.41	1.52	"	20	93,000	68,000	24	66*	.....

\* Houdremont.

**Molybdenum-Vanadium.** A Mo-V steel has much the same nature as the Al-treated Mo steels mentioned in Chap. 9. The Mo gives some depth-hardening properties, and the V gives fine grain. Data for such a steel are given in Fig. 9.

Molybdenum-V steel is used at Watertown Arsenal<sup>15</sup> for the production of centrifugally cast guns, which after suitable heat treatment

and "autofrettage" (See Chap. 7, Vol. I) have strength equal to the strength of forged guns and are cheaper to produce. The composition is around 0.40 C, 0.70 Mn, 0.25 Si, 0.30 Mo, 0.10% V. In sections 1.5- to 2-in. thick, the castings are heated 4 hr. at 1740°, normalized, heated to 1560° for 6 hr., and slow cooled, heated to 1580° for 5 hr., water quenched, and drawn at 1290° for 6 hr.

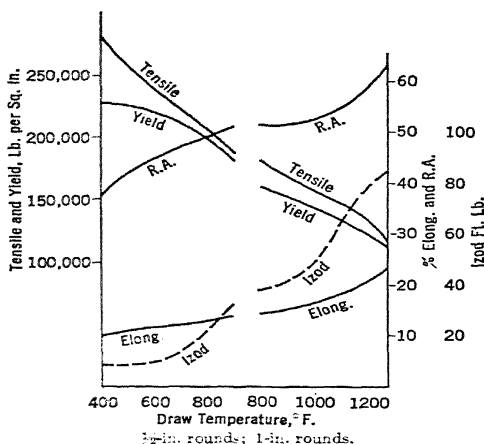


Fig. 9.—Properties of MoV steel, 0.46 C, 0.84 Mn, 0.17 Si, 0.33 Mo, 0.17% V (also 0.08 Cr, 0.09% Ni). Normalized from 1650°, water quenched from 1525°, drawn as shown. (Vanadium Steels and Irons.<sup>7</sup>)

See general comment on property charts, p. 40, chap. 1.

Specimens cut from guns so treated show 95,000–100,000 tensile, 65,000–75,000 yield, 25–30% elong., 55–60% R.A.

A steel of 0.45% C, otherwise similar to the above, in forged sections 3 by 4 in., normalized from 1700°, water quenched from 1570°, and drawn 5 hr. at 850°, showed 126,000–136,000 tensile, 94,000–104,000 yield, 17–21% elong., 57–65% R.A.

Steels for high-temperature service are used which contain the carbide formers Mn, Cr, and Mo, sometimes V in varying proportion, but usually with at least 0.50% Mo. These will be discussed in Chap. 19.

**Summary.** The steels of about 1% Cr are improved in yield and in ductility and impact by addition of V or Mo. The depth hardening is not much affected by the usual amount of V, but the resulting fine grain tends to produce somewhat more shallow hardening. Molybdenum on the other hand, adds to depth hardening. Depth hardening

is not so strong as in commercial Ni-Cr and Ni-Cr-Mo steels, so that the Cr-V and the Cr-Mo steels may be normalized. The X4130 aircraft steel normalizes well in the thin sections used.

Decided resistance to tempering is conferred by Mo. The high draw temperature permissible allows relief of quenching stresses.

Both Cr-Mo and Cr-V steels will stand, and are indeed benefited by, higher quenching temperatures than would be used for plain Cr steels. Manganese may be raised to the alloying level if Cr is lowered. The Mn-Cr-V combination is especially suitable for normalizing. However, in sizes where quenching is effective, quenched and drawn steels of the general class dealt with in this chapter have better properties than when normalized.

Machinability at a high hardness level is considered to be an attribute of Mo-containing steels in general, and notably of Cr-Mo. Blüthgen<sup>10</sup> has presented additional evidence of this.

When machinable hardnesses are sought and the sections are not too large, the Cr-Mo steels show an excellent combination of yield strength and toughness, at low cost compared with more complex and more highly alloyed steels. They may be improved by additions for grain-size control. Steels containing Cr are seldom used as castings.

More highly alloyed combinations of the steels dealt with in this chapter are employed for many special purposes and will be met in later chapters. Detailed information on Cr-V, Cr-Mo, and more complex constructional steels will be found in the "Alloys of Iron and Chromium," Vol. I<sup>1</sup> and in the "Alloys of Iron and Molybdenum."<sup>10</sup>

## BIBLIOGRAPHY

1. KINZEL, A. B., and W. CRAFTS: "The Alloys of Iron and Chromium." New York, McGraw-Hill Book Co., Vol. 1, 1937, 535 pages.
2. GREAVES, R. H.: "Chromium Steels." London, Department of Scientific and Industrial Research, His Majesty's Stationery Office, 1935, 321 pages.
3. JONES, J. A.: Chromium-Copper Structural Steels, *J. Iron Steel Inst.*, v. 121, 1930, pp. 209-224.
4. BUCHHOLTZ, H., and W. KÖSTER: Über die Anlasshärtung kupferlegierten Stahles (Precipitation Hardening of Copper Alloy Steel), *Stahl u. Eisen*, v. 50, 1930, pp. 687-695, discussion, pp. 997-999. Abst.: *Metallurgist*, v. 6, 1930, pp. 87-91. Abst.: *Iron Age*, v. 126, 1930, p. 551.
5. LEQUIS, W., H. BUCHHOLTZ, and E. H. SCHOLZ: Biegeschwungungsfestigkeit und Kerbempfindlichkeit in ihrer Beziehung zu den übrigen Festigkeitseigenschaften bei Stahl (Fatigue Strength and Notch Sensitivity in Their Relationship to the Other Strength Properties of Steel), *Stahl u. Eisen*, v. 53, 1933, pp. 1133-1137.
6. CHANDLER, H. T., and C. N. DAWES: Develops Manganese-Chromium-Vana-

- dium Steel for Automotive Forgings, *Steel*, v. 93, September 25, 1933, pp. 23-26, 30.
7. Vanadium Corporation of America: "Vanadium Steels and Irons." New York, The Author, 1937, 189 pages.
  8. HÖBREMONT, E.: Die Sonderstahlentwicklung unter Berücksichtigung der Rohstofflage (The Development of Special Steels with Reference to Raw Materials), *Stahl u. Eisen*, v. 57, 1937, pp. 480-493.
  9. CAMP, J. M., and C. B. FRANCIS: "The Making, Shaping and Treating of Steel," Pittsburgh, Carnegie Steel Co., 4th Ed., 1925, 1142 pages.
  10. JONES, J. A.: The Influence of Molybdenum on Medium-Carbon Steels Containing Nickel and Chromium, Research Department, Woolwich, *Report* 67, 1926, 59 pages.
  11. GILLET, H. W., and E. L. MACK: "Molybdenum, Cerium, and Related Alloy Steels," New York, Chemical Catalog Co., 1925, 295 pages.
  12. ZEYEN, K. L.: Zur Frage der Schweißempfindlichkeit (The Question of Weldability), *Tech. Mitt.<sup>8</sup> Krupp*, v. 4, 1936, pp. 115-122. Also: *Zd. V., D. u. Ing.*, v. 50, 1936, pp. 969-973.  
ZEYEN, K. L.: Das Schweißen von Stählen höherer Festigkeit (Welding of High Strength Steels), *Stahl u. Eisen*, v. 55, 1935, pp. 901-906.
  13. THEISINGER, W. G.: Heat Effect in Welding—Part III: Certain Alloy Steels, *Iron Age*, v. 139, February 25, 1937, pp. 28-31, 116.
  14. BOLLENBATH, F., and H. CORNELIUS: Zur Frage der Schweißempfindlichkeit von Flugzeugbaustählen (Susceptibility to Cracking of Aircraft Steels), *Arch. f. Eisenforschung*, v. 10, June, 1937, pp. 563-573.
  15. GREGG, J. L.: "The Alloys of Iron and Molybdenum," New York, McGraw-Hill Book Co., 1932, 507 pages.
  16. SISCO, F. T., and D. M. WARNER: Effect of Heat Treatment on the Properties of Chromium-Molybdenum Sheet Steel, *Trans. Am. Soc. Steel Trng.*, v. 14, 1928, pp. 177-192.
  17. JOHNSON, J. B.: Alloy Steel Sheets for Aircraft, *Iron Age*, v. 125, 1930, pp. 502-505.  
JOHNSON, J. B.: Dependence of Aviation on Metallurgy, *Metals & Alloys*, v. 1, 1930, pp. 450-454.
  18. DAWE, C. N.: Chrome-Molybdenum-Steel Applications from the Consumer's Viewpoint, *J. Soc. Automotive Eng.*, v. 10, 1922, pp. 47-50, 62. Condensed: *Iron Age*, v. 109, 1922, pp. 725-728.
  19. BLÜTHGEN, W.: Vergleichende Drehversuche an Chrom-Nickel und Chrom-Molybdenum Baustählen (Comparative Machining Tests on Ni-Cr and Cr-Mo Steels), *Stahl u. Eisen*, v. 58, 1938, pp. 647-650.
  20. BURTON, H. H., and T. F. RUSSELL: Precipitation Hardening in Three Steels Containing Vanadium, *Iron Steel Inst.*, advance copy 4, October, 1938, 18 pages.
  21. Republic Steel Corporation: "Republic Alloy Steels," Cleveland, Ohio, The Author, 1938, 255 pages.
  22. WHITEMORE, H. L., and W. C. BRUGGEMAN: Strength of Welded Joints in Tubular Members in Aircraft, *Report* 348, National Advisory Committee for Aeronautics, 1930, 37 pages.

## CHAPTER 14

### ALLOY STEELS FOR CARBURIZING

**Strength, Hardness and Toughness of the Case.** The object of carburizing and of the heat treatment of carburized articles is to produce a hard, wear-resistant skin. The eutectoid or hypereutectoid steel of the skin is therefore quenched and tempered back only very slightly. The draw temperature is usually from 300 to 350°, say about 325°, so the structure is martensitic, and is ordinarily file hard. The

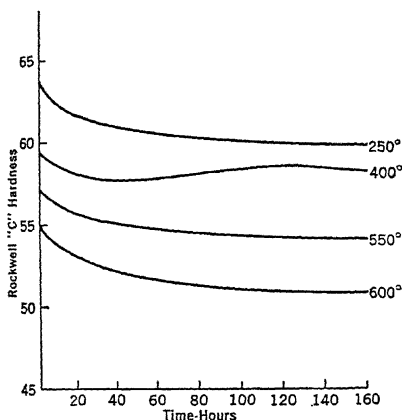


FIG. 1.—Effect of tempering on surface hardness of carburized and quenched 4615. (Nickel Alloy Steels.<sup>22</sup>)

temperature is high enough so that the tetragonal martensite is changed to cubic martensite.

If austenite is retained in the case on quenching, it is not, in most steels at least, broken down into martensite by the usual draw. Figure 1 shows the effect of continued tempering on the surface hardness of S.A.E. 4615, Ni-Mo steel, carburized and oil quenched and tempered 180 hr. at 250°, then the temperature raised, and so on. The curve for 400° shows the decomposition of retained austenite, but that temperature is higher than is ordinarily used.

McQuaid and McMullan<sup>1</sup> studied the surface hardness of carburized S.A.E. 3120 Ni-Cr steel after water quenching, both direct from the pot at 1700°, and also quenched direct, reheated to 1400°,



and again water quenched. The results are shown in Fig. 2. That austenite was retained on the surface in the direct 1700° quench and not decomposed below 350° is plainly shown. However, the retention of austenite did not make the case file soft. File hardness was held, whatever the quench, as long as the tempering temperature did not go above 350°.

Case toughness is thought to be better when some austenite is retained. This condition can be secured with many carburizing steels

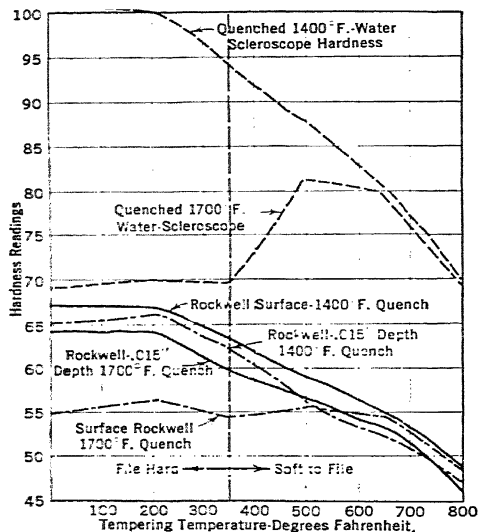


FIG. 2.—Effect of tempering on hardness of carburized 3120. (McQuaid and McMullan.<sup>2</sup>)

on a direct quench, and its presence can be shown by the combination of file hardness with low Rockwell C or scleroscope readings. Unless a considerable amount of austenite is retained, the case hardness, after the normal tempering, will be from 58 to 65 Rockwell C. That is, the tensile strength of the case of a file-hard carburized steel is in the range of 300,000–350,000. This extremely strong material naturally has a tendency to be brittle.

**Necessary Properties of the Core.** The core needs two rather contradictory properties. It needs to be relatively strong in order to support the thin case. (The heavier the case, the more it tends to be self-supporting, but the greater the tendency of the case toward brittleness.) The core also needs to be "tough," but the toughness is not necessarily of the kind that is measurable by the standard tensile, im-

pact, and bend tests. What is really desired is resistance to notch propagation. That is, if the case does crack, the crack should stop when it reaches the core, instead of continuing on through the core. Mere ductility, as ordinarily measured, does not guarantee resistance to notch propagation. For example, if an annealed brass wire that, uncoated, can be twisted about itself is given a thin Cr plating, the plated wire will snap off on being bent through a small angle. If the plating is dissolved off, the bared wire will again stand severe twisting. The crack formed in the brittle coating progresses through the wire, which is tough and ductile by all ordinary criteria.

The desired end is that the carburized piece, core and case together, shall show toughness, as a structure. However, in most instances, any appreciable distortion of a carburized part in service would be a functional failure, so that there is a growing degree of skepticism as to the necessity for large deformation of a carburized part before it will snap. Almen and Boegehold<sup>2</sup> decided that, in rear axle gear service, impact is not much of a factor and that a choice among carburizing steels cannot be made on the basis of impact results alone, even when the impact test is made on an unnotched bar with the case present.

Much of the emphasis laid on the "toughness" of a carburizing steel, put through the carburizing and heat-treating cycle and tested to simulate the core, without the case (or carburized and the case ground off before testing), seems of doubtful import since the core is not actually used without the case and the toughness measured without the case may be quite illusory. The chief need for great resistance to notch propagation is not in normal service, but, instead, in allowing the straightening of warped carburized parts before they are put in service. A better engineering solution is to choose a steel and a treatment that do not cause severe warping.

In the production of carburized gears, where the minimum of grinding after carburizing is desired and where the grinding set-up is adjusted by trial and error to produce the tooth contour that will give quiet gears, it is of the greatest importance that what distortion there is should be uniform from gear to gear, so that they may be machined closely to size before carburizing and yet all clean up with the minimum removal of stock. Spalling of the case is ordinarily ascribable to too sharp a line of demarcation between case and core. In general, it may be overcome by allowing suitable opportunity for diffusion in the carburizing process.

**High-Carbon Steels for Thin Cases.** A re-examination of the engineering requirements for carburized parts is going on, and the carburizing steels of the future may be drawn from a different class than were those of the past.

Most steels for carburizing with heavy cases are low in C, not much over 0.20%, so as to provide for toughness of the core. There is, however, a growing tendency to extend to other uses, beyond the highly stressed gears and steering worms in which such steels have been used, the deep-hardening alloy steels, e.g., S.A.E. 3145 Ni-Cr or 5135 Cr, and other steels of 0.30–0.45% C which are capable of being oil quenched and then tempered at 400–450° to about 275,000 tensile. By previous gas carburizing (see Chap. 10, Vol. I), or cyaniding, such steels may be given a thin, file-hard, wear-resistant skin. Such steels are not necessarily glass brittle; for example, S.A.E. 6140, oil, 400° draw, in small sizes is credited with 10 Izod at 282,000 tensile.

Without carburizing, or with only a flash cyanided skin, such gear steels are generally in the 0.40–0.50% C range. When given a higher C case, a few thousandths of an inch thick, the surface wearing properties are improved. With such a skin, a slightly less hard core will serve as support, and the core toughness can be improved by using lower C, so the trend is toward about 0.30–0.35% C, with fine grain, so as to minimize distortion.

Boegehold and Tobin<sup>2</sup> point out that, beside the saving in carburizing time when only a thin case is required, the alloy steels of 0.30–0.35% C, after suitable normalizing or annealing, are more readily machinable than the 0.20% C grade, which is an added reason for using the higher C content.

**S.A.E. Carburizing Steels.** The S.A.E. low-C carburizing steels and recommended treatments are given in Table I.

The recommendations of the S.A.E. for preliminary treatment before carburizing, for the maximum draw temperature (250° is always the minimum) either after a double or single quench, are listed. These recommendations recognize that the more highly alloyed steels may need normalizing or, in the extreme case, both normalizing and annealing for machinability, and that distortion is minimized by normalizing, before carburizing, at least 50° above the carburizing temperature, on such steels, whereas the slightly alloyed steels do not require it.

The tempering recommendations recognize that it may sometimes be desirable to draw the more strongly hardening steels above the usual 325 or 350° limit, when case toughness rather than extreme hardness is wanted, also that the high quenching temperature when quenching direct will produce greater hardening, and hence may call for a higher draw, than with the lower temperature final quench of a double treatment. Where a direct quench is not considered allowable and so not listed, that fact is shown.

TABLE 1  
S.A.E. CARBURIZING STEELS

S.A.E. No.	C	Range, %				Preliminary Treatment	Quench in	Maximum Draw Temperature	
		Mn	Ni	Cr	Mo			After double Deg.	After direct Deg.
C	1013	a	.....	.....	.....	None required	W	325	325
	1015	a	.....	.....	.....		W	.....	.....
	1020	a	.....	.....	.....		O or W	.....	.....
	X1015	0.70-1.00	.....	.....	.....		O or W	.....	.....
	X1020	0.15-0.25	.....	.....	.....		O or W	.....	.....
Mn	X1314	1.00-1.30	.....	.....	.....	May normalize to minimize distortion and anneal for machinability	O or W	300	300
	X1315	1.30-1.60 <sup>b</sup>	0.40-0.60	.....	.....		O or W	300	300
	2015	a	1.25-1.75	.....	.....		O or W	300	300
Ni	2215	a	3.25-3.75	.....	.....	Normalize and anneal for machinability	O or W	300	300
	2320	a	3.25-3.75	.....	.....		O or W	300	300
	2515	a	4.75-5.25	.....	.....		O	300	300
Ni-Cr	3115	0.10-0.20	1.00-1.50	0.45-0.75	.....	Normalize and anneal for machinability	O or W	300	300
	3120	0.15-0.25	1.00-1.50	0.45-0.75	.....		O or W	300	300
	3215	0.10-0.20	1.50-2.00	0.90-1.25	.....		O	500	No dir.
Ni-Cr	3220	0.15-0.25	1.50-2.00	0.90-1.25	.....	Normalize and anneal for machinability	O	500	No dir.
	3312	0.17 max.	3.25-3.75	1.25-1.75	.....		O	300	400
	3415	0.10-0.20	2.75-3.25	0.60-0.95	.....		O	300	400
Cr-Mo	4120 <sup>a</sup>	0.15-0.25	.....	0.15-0.25	0.15-0.25	Normalize	.....	300	400
	4320	0.15-0.25	1.65-2.00	0.30-0.60	0.20-0.30		O	300	400
	4615	0.10-0.20	1.65-2.00	0.65-0.70	0.20-0.30		O	300	400
Ni-Mo	4620	0.15-0.25	1.65-2.00	.....	0.20-0.30	Normalize and anneal for machinability	O	300	400
	4815	0.10-0.20	0.40-0.60	3.25-3.75	0.20-0.30		O	300	400
	4820	0.15-0.25	3.25-3.75	.....	0.20-0.30		O	300	400
Cr-V	5120	0.15-0.25	.....	0.60-0.90	.....	Normalize	O	300	No dir.
	6115	0.10-0.20	.....	0.80-1.10	.....		O	350	No dir.
	6120	0.15-0.25	.....	0.80-1.10	.....		O	350	No dir.

<sup>a</sup> Mn 0.30-0.60 unless noted. <sup>b</sup> 81 0.75 0.15. <sup>c</sup> Not in official S.A.E. list but commonly so designated.

The carburizing temperature suggested for all steels is 1650–1700°. The usual quenching medium, i.e., oil, water, or either one, according to desire, is also shown in Table I. An exception from the entries shown, is that 1010 and 1015, when the treatment involves reheating to 1650–1700°, may be quenched from that heat in oil or water, but the final quench from 1400–1450° is in water. These quenching medium recommendations merely recognize that, when the steel is so alloyed that the case will harden completely in oil, oil should be used to minimize cracking.

The temperatures of reheating to refine core and case depend on the critical temperature of low C and high C of the given alloy composition. These temperatures are listed in Table II.

TABLE II  
TEMPERATURES FOR HEATING S.A.E. CARBURIZING STEELS

	Heating temp. for core, deg.	Heating temp. for case, deg.
1010	1650–1700	1400–1450
1015		
1020		
X1015		
X1020		
X1314		
X1315		
2015	1650–1700	1375–1425
2115		
2315	1450–1500 core only	1350–1400
2320	1500–1550 prelim. <sup>a</sup>	
2515	1400–1475 core only	1325–1375
	1500–1550 prelim. <sup>a</sup>	
3115	1450–1500 core only	1400–1450 case only 1375–1425 final
3120	1525–1575 prelim.	
3215	1425–1475 core only	1375–1425
3220	1525–1575 prelim.	
3312	1450–1500 core only	1375–1425
	1525–1575 prelim.	
3415	1450–1500 core only	1375–1425
	1500–1550 prelim.	
4320	1475–1525 core only	1425–1475 case only 1375–1425 final
4615	1525–1575 prelim.	
4620	1450–1500 core only	1350–1400
4815		
4820	1550–1650 prelim.	1425–1475
5120	1475–1525 core only	
6115	1600–1650 prelim.	
6120		

<sup>a</sup>ry treatment, when followed by another treatment for the case.

The proper temperatures having been given above, the treatments given below are understood to be carried out at the proper temperature, and the proper quenching medium used as shown in the previous table, with preliminary normalizing or normalizing and annealing as previously specified.

Certain steels are not listed by the S.A.E. as suitable for direct quenching. These are 2515, 3215, 3220, 5120, 6115, and 6120.

In the following table the Roman numerals show the order in which the various treatments are listed for each steel in the "S.A.E. Hand-

TABLE III  
TREATMENTS FOR S.A.E. CARBURIZING STEELS

	Carburize, quench direct, temper	Carburize, quench, reheat for case, quench, temper	Carburize, quench, reheat for core, quench, temper	Carburize, slow cool, reheat for core, quench, temper	Carburize, slow cool, reheat for case, quench, temper	Carburize, slow cool, reheat for prelim. quench, reheat for case, quench, temper	Heat in cyanide or activated bath, to deg., then quench, temper if desired <sup>a</sup>
1010, 1015, 1020, X1015 X1020, X1314, X1315, 2015, 2115		II	.....	III	.....	IV	V 1500-1650 except (2115-1450 1650,
2315 <sup>b</sup> 2320	IV	.....	I	II	.....	III	V 1450-1650
2515	....	.....	I	.....	.....	II	.....
3115 3120	IV	I	I	II	II	III	V 1500-1650
3215, 3720	....	I	I	II	II	III	.....
3312	IV	I	.....	II	II	III	.....
3415 <sup>c</sup> 4320, 4615 4620 4815 <sup>d</sup> , 4820 <sup>d</sup>	IV	I	I	II	II	III	.....
5120 6115 6120	....	I	I	II	II	III	V 1500-1650

<sup>a</sup> Refining heat treatments may be given as for carburized material, to parts cased in activated media.

<sup>b</sup> Single quench can be used on 2315 if maximum core refinement is not needed.

<sup>c</sup> Use oil for direct quenching 3415.

<sup>d</sup> 4815 and 4820 should be fine grained for single quench.

book." When alternatives are shown under the same number, that number appears in two columns. Table I should be consulted as to preliminary normalizing.

Notes in the "S.A.E. Handbook" state that 1112 (0.08–0.16 C, 0.60–0.90 Mn, 0.09–0.13 P, 0.10–0.20% S) may be carburized or cyanided, but 1500–1650° in cyanide or activated bath is given as the preferred method. The higher S (0.20–0.30% S) X1112 steel is not recommended for carburizing or heat treatment.

The 0.50% Ni steel 2015 can be hardened with an oil quench and so may be used to replace C steel when water quenching would have to be used and distortion would result. It is also stated that 2320 seldom requires water quenching. In relation to 2315 it is noted that double heating will produce more distortion than the direct quench. When exceptional core toughness is required in 2515, the C is held to 0.17% max. and the steel then known as 2512.

3230 and 3325 are mentioned as used for parts that must be capable of resisting severe dynamic stresses in service. 3415 is suggested for parts requiring high core strength. 4615 and 4620 are credited with high fatigue resistance and tensile properties, and described as excellent for minimum distortion if fine grained. 4815 and 4820 are said to be used interchangeably with 2515. 5120 is said to be like 2320 and 3120, but not to have so tough a core.

A detailed account of the alternative treatments for each steel is given in the "S.A.E. Handbook." To condense this information, it has been summarized here in tabular form. All entries in Tables I, II, III should be consulted for each steel.

**Range of Core Properties According to Composition of Steel.** The cores of plain C carburizing steels give, on water quenching from 1425° (for the case), about 50,000–95,000 tensile, 50,000–60,000 yield, 20–30% elong., 50–65% R.A.; if quenched from 1625° (for the core), the figures run 105,000, 75,000, 12, 25.

For greatest toughness, steels of 0.10–0.14% C are used, but for good machinability a less gummy steel may be chosen, the C rising to 0.15–0.25%. For the two general types C contents of 0.12 and 0.15% are perhaps the favorites.

The lower alloy carburizing steels, however, after the quench or quenches and draw of the heat treatment after carburizing, give tensile strengths only in the range of 80,000–140,000, so that the case must be thick enough to support itself, and the core's chief purpose is to give toughness.

**Highly Alloyed Steels.** It is possible, at the expense of a considerable alloy content, to secure very high properties in a case-hard-

ened steel. Kallen and Schrader<sup>4</sup> report the following average properties for the core of a case-hardened steel of 0.17 C, 0.30 Si, 0.40 Mn, 4.25 Ni, 1.50 Cr, 0.90% W: 200,000 tensile, 140,000 yield, 14% elong., 55% R.A., 47 mkg./cm.<sup>2</sup> Charpy.

The so-called "Krupp" analysis, like the above but without the W, gives about 150,000 tensile in 2-in. sections. The straight 5% Ni steel, 2515, oil quenched from 1450° shows, in small sizes: 200,000 tensile, 160,000 yield, 12% elong., 50% R.A. By dropping the quenching temperature to 1400°, this steel gives 145,000 tensile, 115,000 yield, 17% elong., 55% R.A.

Such steels, as well as S.A.E. 4615, 1.75 Ni, 0.25% Mo, because of high core strength and strong depth-hardening properties, are used for the most severe service, such as large roller bearings, generally with a rather deep case. 4615 quenched from 1500° after carburizing, slow cooling and reheating, will show about 160,000 tensile with 33 Izod in a 0.4-in. square bar.

**Grain Size.** As was discussed in Chap. 9, Vol. I, the carburizing steels used to be coarse grained, and fine grained "abnormal" steels, with their accompanying tendency toward shallow hardening and propensity toward soft spots under gas bubbles formed in quenching, were avoided. With the realization of the better impact and general toughness of the fine-grained steels, the attitude has been reversed. Hardenability is increased to the necessary degree by more generous use of alloying elements, especially Mn and Mo, instead of using plain C or very highly alloyed steels. With hardenability assured, using a fine-grained, high coarsening temperature, grade, in turn assures toughness.

With steels of high coarsening temperature, the properties on direct quenching are often quite satisfactory. The use of a single quench minimizes distortion, and saves time and money over the heat-treatment methods requiring one or more reheatings.

For these reasons, carburizing steels today are almost universally specified fine grained, i.e., of high coarsening temperature.

**Variation in Properties with Composition and Treatment.** With many possible quenching temperatures, with possible variations in grain size, and with the marked effect of small variation in C content, and of variations in size of bar tested, the core properties of a given alloy carburizing steel after carburizing and heat treatment are not definable without giving all details. Water quenching of 1020, for example, produces variations, as the quenching temperature is raised from 1425° to 1625° of 80,000–105,000 tensile, 50,000–75,000 yield, 30–10% elong., 65–25% R.A.



**Results of Alloying.** Data given by Dawe,<sup>5</sup> unless otherwise , for alloy carburizing steels can be assembled as follows:

Steel	Quenched from, Deg.	In	Tensile	Yield	Elong. %	R.A. %
2115	1600	oil	125,000	100,000	15	45
2115	1600	water	155,000	115,000	10	35
2512	1350	oil or water	110,000	70,000	25	60
2512	1400	oil or water	145,000	115,000	17	55
2512	1450	oil or water	200,000	160,000	12	50
2515	1425	oil	168,000	153,000	14	53 <sup>a</sup>
2315	1450	oil	180,000	91,000	23	57 <sup>a</sup>
3115	1450	oil	109,000	68,000	28	60 <sup>a</sup>
3115	1400	oil or water	125,000	85,000	20	45
3115	1475	oil	165,000	125,000	15	40
3312	1450	oil	210,000	180,000	13	55
3415	1425	oil	140,000	110,000	17	48
4615	1425	oil	105,000	70,000	25	55
4615	1525	water	150,000	110,000	13	50
4615	1460	oil	145,000	92,000	17	47 <sup>b</sup>
4820	1700	oil	140,000	110,000	18	55
4820	1475	oil	135,000	105,000	21	58
4120	Direct quench		136,000	105,000	18	47 <sup>c</sup>
4120	Pot cooled, reheated, quenched		121,000	98,000	23	49 <sup>c</sup>
6115	1700	oil	133,000	95,000	22	62
6115	1625	oil	125,000	90,000	27	60
6115	1475	oil	104,000	61,000	28	58 <sup>d</sup>
6115	1540	oil	109,000	60,500	24	51 <sup>d</sup>
6115	1520	water	138,000	95,500	17	38 <sup>d</sup>

<sup>a</sup> Bethlehem Alloy steels.<sup>6</sup>

<sup>b</sup> Langenberg and McKnight.<sup>7</sup>

<sup>c</sup> Climax Molybdenum Company—private communication.

<sup>d</sup> Vanadium Steels and Irons.<sup>8</sup>

It will be noted that McMullan's data, cited later, do not always agree closely with those listed above.

**Purposes in the Choice of Alloying Elements.** The introduction of alloying elements may be designed to produce a tough case, or a very hard one with excess carbides. As McQuaid and McMullan<sup>1</sup> point out, for resistance to pure abrasion, excess carbides are helpful, though they may give a brittle case; but, to resist battering wear, a case with considerable retained austenite, soft to Rockwell cone and scleroscope hardness tests, but still file hard, is more useful. They find 3120 suitable for a tough case.

The alloying elements are chosen also for the toughness they confer

upon the core. The elements that give the high-yield-strength, low-alloy steels (Chap. 11) have a similar effect here. The two classes of steels are quite similar, save that in the carburizing steels P is kept low, since it would produce brittleness in the high-C case, and Si is held down, since it impedes carburizing. Both classes are low-C steels to which alloying elements are added to give the highest feasible yield strength with the least feasible loss in ductility.

**Sulphur and Phosphorus.** Spalding<sup>9</sup> noted that ordinary screw stock carburized very slowly. To obviate this, free-cutting steels with high Mn are utilized in several commercial steels for carburizing.

For similar reasons the S content of carburizing steel is often thought to be of importance. Bramley, Haywood, Cooper, and Watts<sup>10</sup> state that S has a tremendous depressing effect on the rate of diffusion of C. One of the reasons for apparent differences in carburizing tendencies of steels may be due to their S contents. (Furthermore, differences in rate of decarburization between steels may also be due to S.) Low rates of carburizing, irregular penetration, and poor and irregular hardening on quenching may occur with ordinary screw stock, so that the carburizing grades of high-S screw stock generally have high Mn (e.g., the S.A.E. X1300 series). With Mn around 1 to 1.5%, in an 0.08–0.15% S steel, the S is present, not primarily as FeS, but rather as MnS, a form in which it will not interfere with carburizing; and the excess Mn confers better hardenability, thus compensating for any adverse effects of S. The bad effect of high S is quite generally conceded, but McMullan<sup>11</sup> denies it and ascribes the trouble noted to P rather than to S.

At any rate, S in carburizing compounds is to be avoided. Cases are on record where charred leather gave up S and raised the S content of an ordinary carburizing steel of normal Mn content to 0.55% at the surface, producing a brittle case. There was not enough Mn to combine with this S. The absorption of S was less at 1800° than at 1700°. McMullan,<sup>11</sup> however, insists that sulphides are not disadvantageous to carburizing, and that as much as 0.10% P is without effect on the carburizing process. However, P is held low in carburizing steels, not only because it would produce brittleness in the high-C case, but also because it would be expected to slow down the rate of diffusion.

**Copper.** That Cu in steel does not prevent carburization is interesting when it is recalled that Cu plating is an excellent preventive of carburizing. Epstein and Lorig<sup>12</sup> showed that if a Cu steel is machined to free it from any Cu coat that has been formed by previous preferential oxidation, the mere presence of Cu does not interfere with carburizing.

**Effect of Individual Elements.** An alloying element much used is Ni, since it adds strength and toughness to the ferritic core. It is most used in amounts of 1.25 to 1.75%, though up to 5% may be used when the increased toughness is worth the price. Nickel slightly retards the rate of carburization. As it is not a carbide former, it does not alter the nature of the carbide particles.

Manganese and Cr are carbide formers, tend to accelerate carburization slightly, and to form a harder, more wear-resistant case. To offset the increased hardness and brittleness and to avoid the tendency toward grain growth (especially notable with Mn), they are often accompanied by Ni for toughness or V for grain-growth control. The amount of Cr may vary from 0.50 to 1.25%. Steel of 0.15 C, 1.25 Ni, 0.60% Cr is a standard for high-grade work. Manganese used to be avoided in carburizing steels for fear of brittleness of the case and the danger of grain growth in the core. Now that fine-grained steels with high coarsening temperatures are available, this situation has altered: X1015 with 0.70 to 1.00% and X1314 with 1.00 to 1.30% Mn are standard carburizing steels. McQuaid and McMullan<sup>13</sup> even commented rather favorably on a steel with 3% Mn and 0.30% Mo, grain size 8-9, hardened by air cooling without quenching, since a steel so alloyed is very strongly air hardening.

Molybdenum is a carbide former as well as a ferrite strengthener. It makes for rapid absorption of C. It is popular in carburizing steels, especially in combination with Ni in a steel of about 0.15 C, 1-1.75 Ni, 0.20% Mo, or with Cr in 0.20 C, 0.70 Cr, 0.20% Mo.

**Composition vs. Rate of Carburization.** Some authorities state that Ni slightly reduces and some that it slightly increases depth of penetration. Spalding,<sup>9</sup> who reports the former, found that both Cr and Mo hastened penetration. This is in agreement with Houdremont and Schrader<sup>14</sup> whose data will be referred to later, on Mo but not on Cr. Guthrie<sup>17</sup> gas-carburized low C-Mo and low C-Cr steels under the same conditions and found the same depth of case, but the surface skin of the Mo steel had 0.90% C; that of the Cr steel was highly hypereutectoid.

The C content of the surface of the case will vary, not only with the carburizing temperature, but also with the alloying elements in the steel. Moreover, the rate of diffusion of C through austenite, from the outer case toward the core, will vary with the other elements dissolved in the austenite. However, McQuaid and McMullan<sup>13</sup> tested the depth of hypereutectoid case and the total depth of case for a series of carburizing steels of different analyses, all pot carburized 12 hr. at 1675°. With the exception of a 5% Ni steel, which had a

slightly shallower case with no hypereutectoid zone, the results ranged: hypereutectoid layer 0.009 to 0.018 in., eutectoid and hypereutectoid layer 0.024 to 0.036 in., total depth to original core C content, 0.043 to 0.063 in. But note the results on two steels of the following analyses:

Analysis %	C	Mn	P	S	Si	Ni	Mo	Grain Size	Case Depth, Inches		
									Hyper-eutectoid	Eutectoid and Hyper-eutectoid	Total
7A	0.18	0.59	0.15	0.02	0.16	1.74	0.29	10	0.018	0.036	0.060
7D	0.18	0.53	0.20	0.03	0.23	1.83	0.23	9-10	0.010	0.024	0.048

Thus the variation in two steels to the same chemical specification was as wide as that between several quite different types of steels.

Almen and Boegehold,<sup>2</sup> on steels pot carburized at 1650°, give, for total case depth:

Steel	INCHES	
	4 hr.	8 hr.
1315 Mn	0.046	0.067
2315 Ni	0.041	0.053
2330 Ni	0.035	0.050
2512 Ni	0.038	0.054
3130 Ni-Cr	0.042	.....
4815 Ni-Mo	0.041	0.057
1.75 Mn, 0.23% Mo	0.045	0.060

Rockwell and Downes<sup>16</sup> give, for steels carburized at 1700° pot cooled and oil hardened, for the case:

Steel	Depth by Fracture Inches	
	5 Hr.	9.5 Hr.
1015 C	0.034	0.060
2315 Ni	0.036	0.063
3115 Ni-Cr	0.043	0.064
6120 Cr-V	0.038	0.053
2512 Ni	0.038	0.065

The effect of individual elements was also studied by Houdremont and Schrader<sup>14</sup> by box carburizing in charcoal plus a large amount of  $\text{BaCO}_3$  at  $1675^\circ$ . The pots took 6.5 hr. to come to temperature and were held at temperature for varying lengths of time. For 10 and 30 hr. at heat, they report the following, on steels with 0.15% C plus the alloying elements shown:

Steel	10 Hr.		30 Hr.	
	Max. C %	Case Depth Inches	Max. C %	Case Depth Inches
No alloy	1.10	0.05	1.20	0.11
1.4% Si	0.75	0.035	0.95	0.11
1.7% Al	0.95	0.035	1.25	0.085
3.1% Ni	0.90	0.06	1.00	0.095
0.5% V	1.10	0.06	0.90	0.11
1.5% Co	0.90	0.05	1.10	0.11
1.6% Cu	0.95	0.045	1.10	0.105
1.5% Mn	1.05	0.06	1.10	0.115
1.25% Ti + 0.15% Al	1.25	0.045	1.35	0.08
1.6% W	1.35	0.05	1.40	0.09
0.75% Mo	1.30	0.06	1.40	0.11
1.5% Cr	1.35	0.04	2.60	0.095

The strong graphitizers, Si and Al, interfere with carburizing; Ni, Co, and Cu, also graphitizers, tend to give a lower C case. Vanadium, a strong carbide former, did not build up a high-C surface and the surface C fell on long treatment. McQuaid<sup>17</sup> points out that V steels carburized at low temperature and quenched from the pot tend to build up a high carbide concentration on the surface so that it works better to carburize them at  $1775^\circ$  and get rapid enough diffusion to hold the surface C down. The amount of V in Houdremont and Schrader's steel was, of course, far above the amount normally used in the ordinary carburizing steels.

One and a half per cent Mn apparently hastens diffusion without building up a high-C surface. Of the stronger carbide formers, Mo both builds up a higher C surface and favors diffusion, whereas Cr builds up a high-C surface, remarkably high on Houdremont and Schrader's 30-hr. run, and seems to hinder diffusion.

When an excess of wear-resistant carbides at the surface is desired, Cr and Mo are indicated. For rapid carburizing, Mn and Mo are indicated; for toughness of the core, Ni is helpful. Thus the commercial

alloy carburizing steels chiefly utilize some combination of Ni, Mo, Cr, and Mn, with suitable grain-size control additions.

**Complexity of the Problem of Selection.** The choice of one alloy steel for carburizing over another and the selection of the proper carburizing medium, temperature, and time, and finally of the proper heat treatment, all depend on the service to which the carburized part is to be put. There is little agreement among users. The surface C content, the depth of case, and the gradient within the case can be regulated for the various steels by choice of carburizing medium, temperature, and time. When these desired conditions have been met, one must decide upon the treatment, so as to give both case and core the required balance between hardness and toughness for the alloy steel in question. Jominy and Boegehold,<sup>18</sup> as has been pointed out in Chap. 9, Vol. I, found large variations in hardenability in different lots of carburizing steels of closely agreeing chemical analysis and grain size, which further complicates a choice.

It is generally considered that the most complex heat treatment, i.e., cooling from the carburizing temperature, heating above the critical range of the core, quenching, usually in oil, again heating above the critical range of the case, quenching again, and finally giving a stress-relieving, low-temperature draw, gives the best mechanical properties. But a single reheat and quench from the lower temperature, i.e., primarily for the benefit of the case, are advocated by Richardson<sup>19</sup> for the Ni steels, 2315 and 2512, the core properties being practically the same with the single as with the double quench. McQuaid and McMullan<sup>13</sup> found that 4615, Ni-Mo single quenched from above the upper critical of the core, gave the most satisfactory results of the series they studied, where high surface hardness and high impact resistance were wanted; when high impact resistance without extreme surface hardness was sought, the high Ni 2512 gave the best combination.

**Brittleness.** The impact properties of the toughest carburized steels are low. The core alone, put through the heat-treating cycle, but without being carburized, may show 50 to 100 ft.-lb. Izod. Tested as carburized notched bars, McQuaid and McMullan<sup>13</sup> found that only 1 to 10 ft.-lb. was obtained in many cases, values of 20 ft.-lb. or over being obtained only on certain Ni-Mo steels. Other Ni-Mo steels of similar composition gave 3 to 6 ft.-lb., the difference probably being due to difference in grain size.

McMullan<sup>20</sup> has presented a large amount of data on case and core properties of carburizing steels, all in three conditions of treatment. Carburizing was at 1700°, the steels being either (1) direct quenched,

(2) single treated, i.e., slow cooled, heated and quenched, or (3) double treated, i.e., slow cooled, reheated, quenched, again reheated and quenched. Except 1020, all were quenched in oil. The temperatures used in single and double treatments were:

TEMPERATURE		
Steel	Single Deg.	Double Deg.
1020	1425 water	1625 (oil) 1425 (water)
2315	1475	1550 - 1380
3115	1475	1550 - 1425
4615	1500	1550 - 1400
2512	1450	1550 - 1350
6115	1625	1650 - 1475
Cr-Ni-Mo	1500	.....
Krupp	1450	1550 - 1400
4820	1475	1550 - 1400

Either the single or the double treatment was concluded with a 325° draw.

The compositions and McQuaid-Ehn grain sizes are shown below:

	Grain Size	C %	Mn %	Ni %	Cr %	Mo %	V %
1020	2-7	0.16	0.59	.....	.....	.....	.....
"	2-3	0.19	0.54	.....	.....	.....	.....
2315	7	0.19	0.61	3.62	.....	.....	.....
"	2-3	0.16	0.42	3.60	.....	.....	.....
3115	2-7	0.18	0.48	1.32	0.65	.....	.....
"	2-6	0.15	0.47	1.30	0.59	.....	.....
4615	7-8	0.17	0.54	1.77	.....	0.25	.....
"	3	0.16	0.54	1.82	.....	0.24	.....
2512	6-7	0.15	0.53	5.00	.....	.....	.....
"	5	0.16	0.45	4.94	.....	.....	.....
6115	8	0.18	0.76	.....	0.89	.....	0.18
Cr-Ni-Mo	4	0.16	0.82	1.75	0.80	0.40	.....
"	4	0.25	0.67	1.76	0.80	0.40	.....
Krupp	7	0.08	0.48	3.99	1.48	.....	.....
4820	7	0.18	0.63	3.49	.....	0.24	.....

The properties for the three treatments are shown in the following tables:

## CORE PROPERTIES

287

Steel	Grain Size	Tensile	CORE TENSILE		1700° Oil <sup>a</sup> (Unless Noted) R.A. %	Av. Izod			Av. Rockwell C. of Case	File Test
			Yield <sup>b</sup>	Elong. %		No Case	With Case <sup>c</sup>	With Case Unnotched		
1020	2-7 <sup>d</sup>	102,000	45,000	18	54.5	42	5	9	64	hard
"	2-8 <sup>d</sup>	109,000	40,000	14	39.5	16	4	14	64	"
2315	7	166,500	75,000	14.5	44.7	34	8	25	54	soft skin
"	2-8	155,000	75,000	13.5	45	31	10	30 <sup>e</sup>	55	"
3115	2-7	127,000	60,000	18.5	52	42	5	13	58	hard
"	2-6	108,000	45,000	23.5	63.5	83	7	20 <sup>e</sup>	59	"
4615	7-8	115,000	55,000	25.5	64.5	96	8	18 <sup>e</sup>	59	"
"	3	134,000	55,000	15	55	30	7	15 <sup>e</sup>	58	"
2512	6-7	191,000	110,000	13.5	50	32	8	26	53	soft
"	5	197,000	120,000	14.5	50.5	29	6	20	53	"
6115	8	131,000	70,000	20.5	61.5	80	6	13	56	"
Cr-Ni-Mo	4	168,000	80,000	15.5	56	42	8	18	53	hard, light soft skin
"	4 <sup>f</sup>	225,000	135,000	12.5	47.5	33	6	15	50	"
Krupp	7	157,000	75,000	17.5	65	77	9	33	50	"
4820	7	142,000	65,000	19	55	.....	9	28	59	"

<sup>a</sup> To correspond to pet quenah.<sup>b</sup> Johnson limit.<sup>c</sup> High C case—specimens notched before carburizing.<sup>d</sup> Water.<sup>e</sup> Variable.<sup>f</sup> Machined to size before treating.



CORE PROPERTIES - SINGLE TREATMENT

Steel	Crain Size	Tensile	Yield	Elong, %	R.A., %	Notched		With Case Unnotched Izod	Rockwell C, Case	File Test
						Izod No Case Av.	With Case Av.			
1020	2-7	83,000	30,000	32.5	62	60	3	6	63	hard
"	2-3	92,000	30,000	24.5	45.5	13	2	4	63	"
2315	7	140,000	65,000	18.5	50.5	48	7	40 <sup>a</sup>	58	soft
"	2-3	131,000	50,000	19	50	56	9	18 <sup>a</sup>	58	"
3115	2-7	125,000	40,000	23.5	50.5	49	21 <sup>1/2</sup>	7	62	hard
"	2-6	100,000	30,000	28	59.5	66	3	6	62	"
4615	7-8	114,000	50,000	27.5	62	87	12	16 <sup>a</sup>	61	"
"	3	125,000	55,000	20.5	61	40	4	8 <sup>a</sup>	61	"
2512	6-7	187,000	105,000	15.5	54.5	32	7	30 <sup>a</sup>	56	soft
"	5	197,000	115,000	15	56.5	27	4	15	57	"
6115	8	127,000	65,000	23.5	59.5	81	5	34 <sup>a</sup>	62	hard, soft skin
Cr-Ni-Mo	4	150,000	65,000	15.5	62	58	8	14	61	"
"	4 <sup>b</sup>	231,000	135,000	14	49.5	33	4	9	61	"
Krupp	7	151,000	70,000	18	64.5	65	5	13 <sup>a</sup>	61	soft skin
4820	7	135,000	60,000	21	59.5	61	6	25	60	hard

<sup>a</sup> Results variable.<sup>b</sup> Machined to size before treatment.

CORE PROPERTIES—DOUBLE TREATMENT

Steel	Grain Size	Tensile	Yield	Elong. %	R. A. %	Notched No Case	Izod With Case	Unnotched With Case	Rockwell C. Case	File Test
1020	2-7	82,000	35,000	34	69	98	4½	10	65	hard
"	2-3	86,000	35,000	30	59	51	4	14 <sup>a</sup>	65	"
2315	7	125,000	35,000	29.5	54	57	8	28 <sup>a</sup>	56	fairly hard, soft skin
"	2-3	113,000	35,000	28	57	76	11	30 <sup>a</sup>	57	"
3115	2-7	117,000	40,000	28	52	59	6	13 <sup>a</sup>	59	hard, thin soft skin
"	2-6	100,000	30,000	32.5	60.5	79	8	20 <sup>a</sup>	60	"
4615	7-8	108,000	40,000	32	64	93	9	20 <sup>a</sup>	60	"
"	3	109,000	40,000	29.5	56.5	71	7	10 <sup>a</sup>	59	"
2512	6-7	153,000	60,000	15.5	55	42	8	26	53	soft, deep skin
"	5	147,000	55,000	17.5	56	54	7	20	54	"
6115	8	112,000	40,000	27.5	56.5	56	6	13 <sup>a</sup>	57	hard, below soft skin
Cr-Ni-Mo	4	.....	.....	.....	.....	.....	8	18	52	hard, below thin skin
"	4	.....	.....	.....	.....	.....	5	14	50	"
Krupp	7	156,000	75,000	19	65.5	67	8	33	50	"
4820	7	134,000	60,000	21	60.5	60	9	28	58	"

<sup>a</sup> Results variable.

All these specimens were carburized to give the usual high-C case. Impact tests were also made on specimens with low-C cases, not file hard. With such cases the core toughness was more nearly approached.

Average impact results for all case depths, and C contents are lumped together for six of the steels in Fig. 3. By comparing this figure with the data of the tables, for specimens without cases, it will

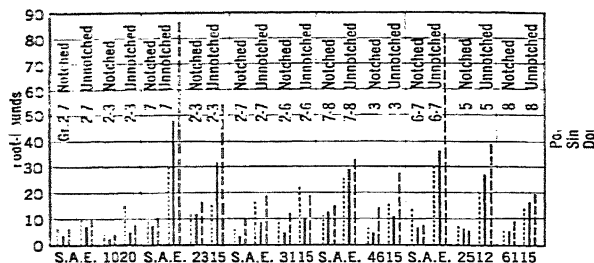


FIG. 3.—Average impact results on case-hardened steels.

The grain size shown is by the McQuaid-Ehn test. (McMullan,<sup>17</sup>

be seen that the toughness of the core itself is *not* necessarily closely reflected in the toughness of the carburized part.

Unnotched impact specimens are often used to evaluate the toughness of the core, since, at the low draws used to maintain file hardness of the case, the impact resistance is so low that the differentiation between steels is none too clear with notched bars. Impact results on unnotched case-hardened bars are usually quite variable.

	C %	Mn %	Ni %	Mo %	Range of Impact Tests, Unnotched Bars, ft.-lb.
2315	0.15-0.18	0.40-0.55	3.60-3.70	.....	90-152
2512	0.12	0.47	4.90	.....	56-119
4815	0.15	0.49	3.44	0.34	55-149
1315	0.14	2.14	.....	.....	61 <sup>1/2</sup> -20 <sup>a</sup> , 90 <sup>b</sup>
....	0.17	1.75	.....	(coarse grain) 0.23 (fine grain)	10-25 <sup>c</sup> , 31-46 <sup>d</sup> , 61-90 <sup>e</sup>

<sup>a</sup> Direct quench, slow cool, reheat 1475°, quench; or cool direct to 1375° and quench from that temperature.

<sup>b</sup> Direct quench, reheat 1400°, quench.

<sup>c</sup> Cooled direct to 1375°, quenched from that temperature or slow cooled, reheated 1475°, quenched.

<sup>d</sup> Quench

<sup>e</sup> Direct

Almen and Boegehold,<sup>2</sup> using unnotched impact specimens, studied a series of steels, carburized at 1650° for 4 and 8 hr., and tested with the case removed from all but one face of the impact bar. The steels were given a variety of heat treatments, i.e., direct quench, slow cooled, reheated and quenched, reheated to and quenched from a range of temperatures. Although the original should be consulted for the data for each heat treatment, the ranges for the lower C steels are shown in the table on page 290. All specimens were drawn at 300°.

When 0.30% C steels were used, the following was obtained:

	C %	Mn %	Ni %	Cr %	Unnotched Izod
2330	0.31	0.70	3.31	....	12-47 <sup>a</sup>
3130	0.31	0.64	1.14	0.70	10-14

<sup>a</sup> These extreme results were on duplicates, quenched direct, reheated 1330°.

**Distortion.** Although pot cooling, followed by double quenching from the temperatures most suitable, respectively, for the core and the case, should give the best mechanical properties, it must be remembered that each heating and cooling increases the tendency toward distortion and warping.

A single quench from the carburizing temperature, with the carburizing carefully controlled as to temperature and time, gives less distortion than do some of the more complex treatments. To prevent grain growth in carburizing, a fine-grained (high-coarsening temperature) steel, alloyed to produce sufficient depth hardening to offset the use of a fine-grained steel, and to give suitable C content in the case and suitable toughness in the core, is required for quenching from the carburizing temperature.

The Ni-Mo 4615 is often advocated as supreme in meeting these requirements. More recently the cheaper 4120, a low-Cr, low-Mo steel of about 0.20 C, 0.40 Mn, 0.60 Cr, 0.20% Mo has been utilized, even though, on the basis of earlier experience with higher Cr, higher Mo steels, the belief has been widespread that the Cr-Mo steels were very subject to distortion. This was because, with the strong depth-hardening effect of the high Cr and Mo, the core was not tough enough on a single quench and double quenching had to be resorted to, which resulted in distortion. By holding the alloying elements low, so that the core is tough, and by using a controlled-grain steel of high-coarsening temperature to aid in toughness and permit heating to a high temperature without grain growth, this once-reviled type of steel, when properly

normalized before carburizing, is found to be satisfactorily free from distortion on single quench, and the wear-resistant Cr carbide in the case gives it some advantage.

Schrader<sup>21</sup> studied a steel of the older type, with high Mn and Cr, 0.22 C, 0.33 Si, 1.10 Mn, 1.25 Cr, 0.24% Mo, and concluded that it should be quenched from 1510–1580°. If so treated, and if a mild carburizer is used so that too high a C content is not built up in the outer surface of the case, he appraised it as about the equivalent of a 1.0 Cr, 4.5% Ni steel.

**Summary.** The general principles of alloying for depth hardening and of grain-size control to permit high quenching temperatures are especially applicable to carburizing steels. Fine-grained steels are generally specified. Case hardness, type of carbides in the case, core toughness, resistance to notch propagation, case thickness and need for support of the case, are all matters of varying importance, depending on the service in hand. The choice of composition and grain size of steel, of carburizing time and temperature, and of the type of heat treatment is a matter for engineering judgment.

Freedom from distortion is often highly important, and, since the fewer the heatings and quenchings the less the distortion, steels that give satisfactory case and core properties on a single quench are deservedly popular. Misapprehension as to the real toughness of the carburized part may arise if only the evidence is considered from tests on the core without the case. Thinner cases, supported by harder cores than have been used in general practice, are becoming more popular, so that oil-hardening alloy steels of around 0.35% C, so cased, need consideration along with the older, heavily cased, 0.10–0.20% C steels.

#### BIBLIOGRAPHY

1. McQUAD, H. W., and O. W. McMULLAN: Relation of Structure to Surface Hardness of a Case Hardened Steel, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 584–600.
2. ALMEN, J. O., and A. L. BOEGEHOLD: Rear Axle Gears: Factors Which Influence Their Life, *Proc. Am. Soc. Test. Mat.*, v. 35, part 2, 1935, pp. 99–146.
3. BOEGEHOLD, A. L., and C. J. TOBIN: Factors Governing Selection of Type of Carburized Case, *Am. Soc. Metals*, "Carburizing Symposium," October 18 to 22, 1937, pp. 67–88.
4. KALLEN, H., and H. SCHRADER: Über Verfahren zur Erhöhung des Oberflächenverschleisswiderstandes von Kurbelwellen (On Process for Increasing the Surface Wear Resistance of Crankshafts), *Tech. Mitt. Krupp*, v. 1, 1933, pp. 58–65.
5. DAWE, C. N.: "The Steel (Physical Property) Atlas." Cleveland, Am. Soc. Metals, 1936, 87 pages.

6. Bethlehem Steel Company: "Bethlehem Alloy and Special Steels." Bethlehem, Pa., Bethlehem Steel Company, 1935, 375 pages.
7. LANGENBERG, F. C., and C. McKNIGHT: Alloy Bearings Applied to Trains, *Iron Age*, v. 121, 1928, pp. 130-131.
8. Vanadium Corporation of America: "Vanadium Steels and Irons." New York, The Author, 1937, 189 pages.
9. SPALDING, S. C.: A Comparison of the Rate of Penetration of Carbon into Various Commercial Steels in Use for Case Carburizing, *Trans. Am. Soc. Steel Treat.*, v. 2, 1921-1922, pp. 950-976.
10. BRAMLEY, A., F. W. HAYWOOD, A. T. COOPER, and J. T. WATTS: The Diffusion of Non-Metallic Elements in Iron and Steel, *Trans. Faraday Soc.*, v. 31, 1935, pp. 707-734.
11. McMULLAN, O. W.: Steels Used in the Carburizing Process, *Trans. Am. Soc. Metals*, v. 26, 1938, pp. 546-567.
12. EPSTEIN, S., and C. H. LORIG: Note on the Carburizing of Copper Steel, *Metals & Alloys*, v. 6, 1935, pp. 91-92.
13. McQUAID, H. W., and O. W. McMULLAN: Selection of Case Hardening Steels for Highly Stressed Gears, *Trans. Am. Soc. Steel Treat.*, v. 16, 1929, pp. 860-892.
14. HOUDREMONT, E., and H. SCHRADER: Einfluss der Legierungselemente auf das Verhalten von Stählen bei der Einsatzhärtung (Influence of Alloying Elements on the Behavior of Steels during Casehardening), *Archiv f. Eisenhüttenw.*, v. 8, 1935, pp. 445-459.
15. GUTHRIE, R. G.: Discussion, *Trans. Am. Soc. Steel Treat.*, v. 14, 1928, pp. 827-828.
16. ROCKWELL, S. P., and F. DOWNES: What Is Case Depth? *Ibid.*, v. 5, 1924, pp. 285-301.
17. McQUAID, H. W.: The Importance of Aluminum Additions in Modern Commercial Steels, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 797-838.
18. JOMINY, W. E., and A. L. BOEGEHOLD: A Hardenability Test for Carburizing Steel, *Trans. Am. Soc. Metals*, v. 26, 1938, pp. 574-606.
19. RICHARDSON, S. A.: Discussion, *Trans. Am. Soc. Steel Treat.*, v. 14, 1928, pp. 854-855.
20. McMULLAN, O. W.: Physical Properties of Case Hardened Steels, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 319-381.
21. SCHRADER, H.: Beeinflussung der Zähigkeit von Einsatzstählen durch Herstellungsart und Wärmebehandlung (Influence of the Toughness of Case-hardening Steels by Production Methods and Heat Treatment), *Stahl u. Eisen*, v. 56, 1936, pp. 1201-1210.
22. FRENCH, H. J., and J. W. SANDS (editors): "Nickel Alloy Steels." New York, International Nickel Co., Inc., 1934, 7 sections.

## CHAPTER 15

### ALLOY STEELS FOR NITRIDING

**Nitrided Steels as Constructional Steels.** In a sense the nitriding steels are "special-purpose" steels, since they are selected for their suitability to absorb N. Yet the necessity of supporting the thin nitride case with a sufficiently hard core and the fact that nitrided parts are used more or less interchangeably with carburized ones (and carburized steels are certainly classed as constructional steels) make consideration of the nitriding steels here rather than in the next section of this volume a bit more logical.

**Support Required for the Nitrided Case.** The nitrided case, as has been pointed out in Chap. 13, Vol. I, in the section on the nitriding process, is very hard and brittle. Because a thick case would be more likely to spall off and because of the long treatment necessary to produce even a thin case, the nitrided layer is very thin. Hence it requires support, much as do the very thin cases on cyanided or gas carburized gears that carry only a surface film which were mentioned in the last chapter. The nitriding steels must then be chosen so that after prior heat treatment the 950°, or thereabout, nitriding treatment will not soften them too much. There is no heat treatment after the nitriding process.

**Preparatory Heat Treatment.** Preparation for nitriding requires quenching, followed by tempering to sorbite, with the avoidance of (or the removal of) any decarburized skin, as has been discussed in Chap. 13, Vol. I. The tempering temperature should be above that used in the nitriding process, and should be held for a long enough time to release internal stress and prevent warping in nitriding. It will be recalled that Mo steels are resistant to tempering, and Mo is present in most nitriding steels.

**Composition of Steels for Nitriding.** The amounts of alloying elements are chosen with an eye both to proper nitride skin formation and to proper depth hardening and toughness of the cores. Ordinarily a steel of around 0.35% C, high in Al, and quite heavily alloyed with depth-hardening elements, is used.

Although it is possible to nitride a plain C steel, the Fe-nitride sur-

face is too brittle and not sufficiently hard. Nitride-forming elements, notably Al and Cr, sometimes V, are employed in order to get nitriding in a reasonable time and to produce the type of nitrides that give a very hard, and not too brittle, skin.

The nitriding temperature is one prone to develop temper brittleness on slow-cooled, temper-brittle steels; and the Cr steels are likely to be temper brittle. Since Mo is a specific against temper brittleness, Mo appears in most nitriding steels.

Commonly used nitriding steels fall into three main groups:

	Group I Cr-Al-Mo		Group II Al-Mo	III Cr-V
%	N125	N135	N230	N630
C	0.20-0.30	0.30-0.40	0.25-0.35	0.24-0.35
Mn	0.40-0.60	0.40-0.60	0.40-0.60	0.40-0.60
Si	0.20-0.30	0.20-0.30	0.20-0.30	0.20-0.30
Al	0.90-1.40	0.90-1.40	1.00-1.50	.....
Cr	0.90-1.40	0.90-1.40	.....	1.40-1.60
Mo	0.15-0.25	0.15-0.25	0.60-1.00	.....
V	.....	0.05 <sup>a</sup>	.....	0.45 min.

<sup>a</sup> If desired.

The C content is sometimes raised to 0.45 or 0.55% in order to secure greater core strength, and the Cr content may be raised over that in N125, e.g., to 1.50-2.00%, to secure a still harder case.

A somewhat softer but tougher case is obtained by raising the Cr to 1.75-2.25% and decreasing the Al to 0.35%.

\* **Steels without Aluminum.** Aluminum can be dispensed with, as is shown above for N630, i.e., Cr-V steels may be used. The 0.50% V in N630 may be replaced by about 0.20 Mo + 0.15% V and the Cr may be raised to around 2%.

A still higher content of alloying elements, with the C correspondingly decreased, is met in a steel of 0.15 C, 0.45 Mn, 0.25 Si, 2.75 Cr, 0.50 Mo, 0.25% V, which, normalized from 1750°, water quenched from 1700°, gives: <sup>1</sup>

Draw	Tensile	Yield	Elong. %	R.A. %	Izod
950°	180,000	152,000	15	58	33
1050°	175,000	150,000	16	60	36
1150°	160,000	145,000	17.5	64	55
1250°	126,000	118,000	19	68	90



Such a steel gives a nitride case with a surface hardness of about 900 Vickers-Brinell against about 1100 for the ordinary Al-Cr-Mo steels, but the case is tougher. Still further increase of Cr to 4.00 to 6.00% in such a steel gives a nitriding composition, which, save for the V, is one much in use, un-nitrided, for high temperature service as oil still tubes. With 6% Cr the surface hardness, nitrided, may rise to 1100. Various other Cr-Mo-V combinations may be utilized.

Some nitriding steels commonly used do not compare unfavorably in core properties with those of many constructional steels, so that any of these steels might be considered for use in the un-nitrided condition. However, steels high in Al are prone to surface defects that have to be ground out of the billets. The care required in manufacture raises their cost above that due to their alloy content alone. Hence the N135 type is almost solely a nitriding steel.

**Special Nitriding Steels.** Additional steels have been evolved to meet special requirements. Other elements than those commonly used might have utility. For example, it is stated<sup>2</sup> that Ta increases the depth of penetration in nitriding, and Zr is suggested<sup>3</sup> in a steel of 1.00-1.50 Cr, 0.60-0.90 Si, 1.00-1.25 Mn, 0.10-0.30% Zr. Steels having a high core hardness are desirable whenever the case is required to resist high compressive stresses such as are present in drawing dies and similar applications.

A composition that shows a Brinell hardness of 500 for the core after nitriding is as follows:

C 0.75, Cr 1.25, W 1.00, Al 1.00, V 0.15, and Mo 0.75%.

This steel is oil quenched from 1650° to 1675° and then tempered at 950°. It should be nitrided at 950° in order to retain maximum core hardness. The nitriding time will vary generally from 48 to 72 hr.

Steels exhibiting secondary hardness can be surface-hardened by nitriding. A high-C, high-Cr steel of the following approximate composition is being used where a high core hardness is desired: C 1.50, Cr 12.00, Mo 1.00, and V 1.00%. High-speed steel may also be given a nitride case, as may the Cr-Ni-W valve steels.

**Nickel Nitriding Steels.** French and Homerberg<sup>4</sup> have studied the rôle of Ni in nitriding steels. Its effects are to strengthen and toughen the case with moderate decrease in case hardness. It strengthens and hardens the core and thereby provides better support to the case. It develops precipitation hardening in steels containing appreciable proportions of Al, by means of which articles may automatically acquire higher core strength during nitriding.

**Machinability.** A high-S nitriding steel<sup>5</sup> can be used to advantage whenever free machining is highly desirable and when maximum im-

fact properties are unnecessary. Figure 1 shows the physical properties of high- and low-S nitriding steels.

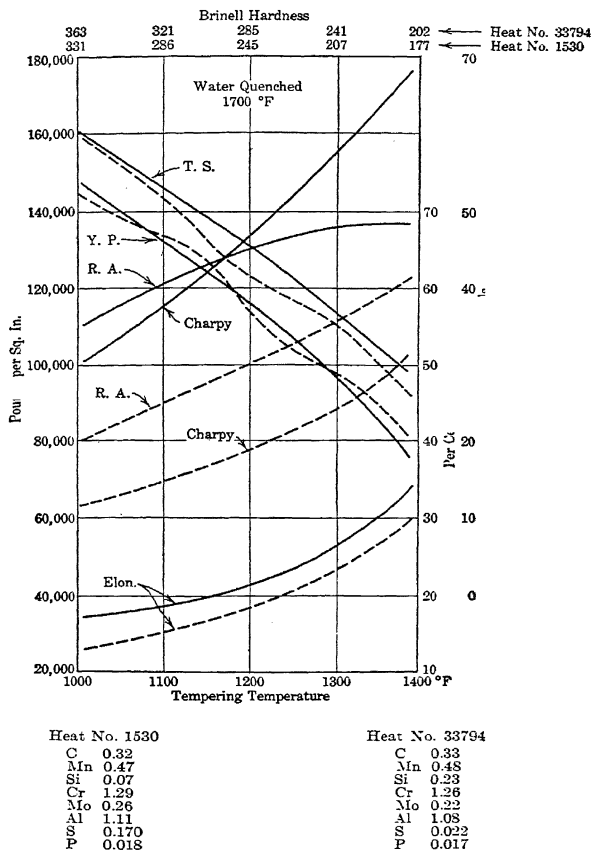


FIG. 1.—Physical properties of high and low sulphur nitriding steels.

See general comment on property chart, p. 40, chap. 1.

The high-S steel machines very readily with a S-base cutting oil. Selenium is also used for machinability, a composition of this type being 0.30–0.40 C, 0.50–1.10 Mn, 0.30 Si max., 0.75–1.50 Al, 0.15–0.25 Mo, 0.15–0.25% Se.

**Properties of Nitriding Steels.** The properties of several nitriding steels (without the nitride case) are shown in Tables I-V.

TABLE I  
PROPERTIES OF N 125

Composition: C 0.23, Mn 0.51, Si 0.20, Cr 1.58, Al 1.24, Mo 0.20, P 0.011, S 0.011%  
Quenched in oil from 1750°. Held at temperature 30 min.

Tempering Temp.	Yield	Tensile	Elong. %	R.A. %	Charpy Impact, ft.-lb.	Brinell Hardness Number
800	156,500	178,500	12	47	20.5	400
900	143,500	170,000	16	49	25.5	365
1000	133,500	160,000	17	54	30.0	340
1100	117,500	134,500	18	61	48.0	290
1200	103,500	122,000	21.5	67	59.0	255
1300	85,500	102,500	27	72	69.0	230
1400	75,000	90,500	32	73	73.0	190
Annealed	60,000	80,000	34	70.5	48.0	155

TABLE II  
PROPERTIES OF N 135

Composition: C 0.36, Mn 0.51, Si 0.27, Cr 1.49, Al 1.23, Mo 0.18, S 0.010, P 0.013%  
Quenched in oil from 1650°\*.

Tempering Temp. deg.	Yield	Tensile	Elong. %		Charpy Impact, ft.-lb.	Brinell Hardness Number
800	180,000	224,500	11	36	12	445
900	165,000	206,500	11.5	37.5	15	415
1000	158,000	182,500	15	50	22	365
1100	137,500	156,000	16.5	57	35	330
1200	120,000	138,000	20	60	44.0	285
1300	103,500	121,000	23	62.5	55.5	225
1400	80,500	104,500	28	59	54.5	200
Annealed at 1450	69,000	95,000	30	67.5	31.5	185

\* In general the present practice is to quench in water from 1700°. The size effect in such steels is shown in Fig. 2.

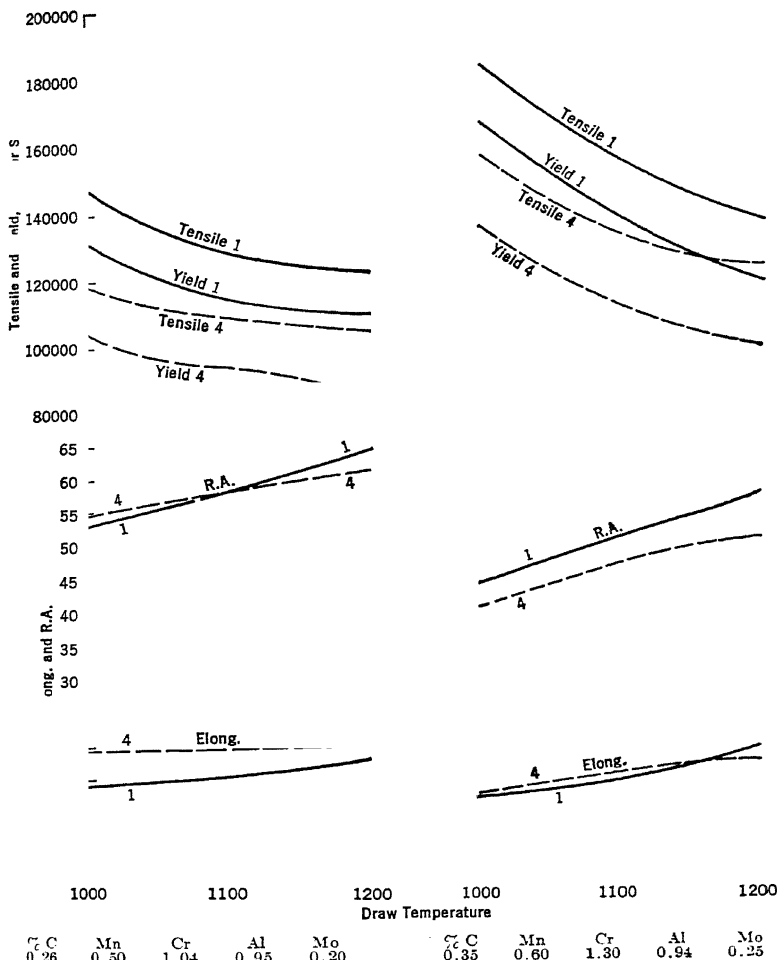


FIG. 2.—Size effect in nitriding

Solid lines, 1-in. diam. bars. Dashed lines, 4-in. diam.

TABLE III

PROPERTIES OF STEEL LIKE N 135, BUT OF HIGHER C

Composition: 0.55 C, 0.55 Mn, 0.15 Si, 1.10 Cr, 1.14 Al, 0.20% Mo.  
Oil quenched from 1550°\*.

Tempering Temp. deg.	Tensile	Yield	Elong. %	R.A. %	Izod	Brinell
1000	206,000	182,500	11	35.5	18.5	450
1100	175,000	160,500	13.5	43.5	28.5	390
1200	158,500	137,000	16	49	40.5	330
1300	130,000	106,500	21	56	59	280
1375	116,500	93,500	23	56	63.5	250

\* 1-in. rounds. Results from water quenching are practically identical in this size. Water quenching is recommended for 2-in. or larger sections.

TABLE IV

PROPERTIES OF N 225\* (AL-FREE)  
Quenched in oil from 1650°

Tempering Temp. deg. for 1 hr.	Tensile	Yield	Elong. %	R.A. %	Brinell Hardness
800	161,500	127,000	12.5	45	320
900	156,000	118,000	14.0	48.0	300
1000	147,000	105,000	18.5	53.5	285
1100	140,500	102,000	22.0	57.0	270
1200	121,500	84,000	25.0	59.5	250
1300	102,500	74,000	28.0	62.5	230
1400	97,500	68,500	31.5	63.5	215
Annealed	89,000	65,500	29.5	61.5	195

\* The C range of this steel is now 0.25-0.35 instead of 0.20-0.30.

TABLE V  
PROPERTIES OF N 630

Composition: 0.33 C, 0.55 Mn, 0.25 Si, 1.37 Cr, 0.50% V.  
Quenched in oil from 1650°.

Tempering Temp. deg. for 1 hr.	Tensile	Yield	Elong. %	R.A. %	Charpy	Brinell
1000	166,500	144,000	15	62	24.5	335
1100	163,000	145,000	17	60	27.5	330
1200	141,500	134,500	21.5	64.5	57.5	300
1300	103,500	97,500	23	75	71.5	210
1400	94,500	77,500	30	75.5	71.5	200

Further data will be found in the section on nitriding, by Homerberg, in the "A. S. M. Handbook"; also in an article <sup>6</sup> by him.

## BIBLIOGRAPHY

1. Vanadium Corporation of America: "Vanadium Steels and Irons." New York. Author, 1937. 189 pages.
2. GENDERS, R.: Tantalum and Niobium, *Sand. Clays. Minerals*, v. 3, April, 1938, pp. 181-185. See also BURDEN, W. M.: Alloy Steels, U. S. Patent 2,103,610, December 28, 1937.
3. KINZEL, A. B.: Nitride-hardened Steel Article, U. S. Patent 2,114,802, April 19, 1938.
4. FRENCH, H. J., and V. O. HOMERBERG: The Rôle of Nickel in Nitriding Steels, *Trans. Am. Soc. Steel Treat.*, v. 20, 1932, pp. 481-506.
5. HOMERBERG, V. O.: High-Sulphur Nitralloy for Free Machining, *Iron Age*, v. 130, 1932, pp. 101-102.
6. HOMERBERG, V. O.: Nitriding, *Iron Age*, v. 138, October 15, 1936, pp. 49-52, 54, 56, 61-62, 64, 66, 98.

## CHAPTER 16

### SPRING STEELS

**Strength and Requirements of Springs.** Spring steels are an excellent example of the choice of compositions being governed far more by minor peculiarities than by the properties measured by ordinary tensile tests.

A spring is a highly stressed part, which must flex repeatedly without showing permanent set. The first requisite is a high yield strength, even though ductility has to be sacrificed to secure it. Steel springs for severe duty are always heat treated: in small sections they may be quenched and tempered before forming or coiling cold, but more commonly they are hot formed (sometimes cold formed), then quenched, and tempered, but not tempered far past the stage of secondary troostite. Quenching is ordinarily in oil.

Tempering is for stress relief, for raising the yield strength (see Chap. 7, Vol. I), and for developing such toughness as is compatible with the yield strength needed. The yield strength sought is usually in the range 175,000–200,000, and the tempering temperatures are therefore low, but definitely higher than for parts (e.g., carburized steels) which are to be file hard and wear resistant.

**Use of Carbon Steels.** The sections are usually rather small. If the steel will harden throughout in the section used, there is little need for alloys; thus watch springs are made of C steel. In automobile leaf springs C steel may not harden clear through on oil quenching and there is danger of cracking in water quenching. In order to get high hardenability of C steel (see Chap. 6, Vol. I) the eutectoid composition may be approached, but the C content is usually held somewhat below eutectoid.

**Use of Alloy Steels.** When C steel will not serve, the C is dropped to the 0.40–0.60% range and some element or elements added that will confer depth hardening, with a preference shown for alloying elements that require elevated tempering temperatures so as to allow as much stress relief as possible.

**Grain Size.** The high yield strength is not compatible with great toughness, so embrittling effects must be avoided, which means chiefly that steels and heat treatments should be chosen to avoid grain growth. The cost factor enters heavily, so the cheaper alloying elements Si and Mn find large use, with considerable use also of Cr–V, and of Mo plus an Al addition, for grain-size control.

**A. S. M. Spring Steels and Treatments for Large Hot-Formed Springs.** Steels listed in the "A. S. M. Handbook" for large, hot-formed flat springs,  $\frac{3}{16}$  in. and over in thickness, or hot coiled helical springs from  $\frac{1}{2}$ -in. diam. bar or over are:

For	C	Si	Mn	Cr	V Min.	Normalizing or Forming deg.	Quench in Oil from deg.	Temper at deg.	To Brinell
Flat, Low C.....	0.70-0.80	.....	0.50-0.60	.....	.....	.....	1450-1475	700-850	352-415
Flat Coil, C-S, A. E. 1095.	0.90-1.05	.....	0.25-0.50	.....	{.....}	1575-1600 1575-1600	1500-1550 1500-1550	850-1050 750-1050	352-415 352-415
Flat Coil, C-Si.....	0.90-1.05	0.25-0.50	0.25-0.50	.....	{.....}	1600-1625 1575-1600	1525-1575 1525-1575	850-1050 750-1050	352-415 352-415
Flat and Coil, Si-C-V.....	0.88-0.98	0.50-0.75	0.45-0.65	.....	0.15	1625-1650	1600-1625	750-1050	363-429
Flat Coil, Cr-V.....	0.45-0.55 0.45-0.55	..... .....	0.70-0.90 0.70-0.90	1.00-1.20 0.90-1.15	0.15 0.15	1600-1625 1600-1625	1600-1625 1600-1625	850-1050 850-1050	363-429 363-429
Flat, Si-Mn.....	0.50-0.60	1.80-2.20	0.60-0.80	.....	.....	1600-1625	1575-1600	850-1050	363-429
Coil, Si-Mn, S.A.E. 9255.	0.50-0.60	1.80-2.20	0.60-0.90	.....	.....	1600-1625	1575-1600	850-1050	363-429



**A. S. M. Spring Steels and Treatment for Small, Cold-formed Springs.** Omitting the cold-rolled or drawn, or previously heat-treated ("pre-tempered" spring stock, which receives only a stress-relieving or a "setting" draw after cold-forming into springs, the annealed steels for cold-formed and then quenched and tempered flat springs from stock under  $\frac{3}{16}$  in. thick, or helical springs from wire  $\frac{1}{2}$  in. or less in diam., listed by the A. S. M. are:

For	Mn	Cr	V Min.	Quench in Oil from deg.	Temper at deg.	To Rockwell C	
						Flat	Coil
Flat and Coil	0.55-0.70	0.90-1.20	.....	1450-1525	700-850	35-40	40-45
Flat and Coil	0.80-0.90	0.30-0.60	.....	1425-1500	800-850	35-40	40-45
Coil	0.90-1.05	0.30-0.50	.....	1450-1525	800-900	35-40	40-45
Flat and Coil	0.45-0.55	0.50-0.80	-1.10 . 0.15	1575-1625	850-950	35-40	40-45

In the above treatments the quenching oil is kept at 100-140°.

**A. S. T. M. Spring Steel Specifications.** Carbon steel springs are used in large sizes in railway service, made from bars as large as  $1\frac{1}{16}$ -in. diam., even though C steel will not harden throughout in such a large size. The compositions are, for elliptical and helical springs, 0.90-1.10 C, for helical only, 0.95-1.15% C, both with 0.50% Mn max. (A. S. T. M. Spec. A-14-27), or the same with Si specified at 0.25-0.50% (A. S. T. M. A68-27). Another specification (A. S. T. M. A 125-33) reads 0.90-1.05 C, 0.25-0.50 Mn, 0.15% Si min. Still another (A 58-27) for vehicle and general-purpose springs calls for 0.85-1.05, or 0.90-1.05% C, both with 0.25-0.50% Mn. The usual Si-Mn (A 59-27) and Cr-V (A 60-27) steels are also used.

**Other Spring Steels.** Among other spring steels commercially produced, for large springs besides the Si-Mn, C-V, Cr-V, and C-Mo steels mentioned below, are:

	Mn	Si	Cr	Mo
C-Cr	0.45-0.55	0.70-1.00	0.10-0.25	1.00-1.30
Cr-Mo	0.45-0.55	0.60-0.90	0.15-0.30	0.80-1.10
Cr-Si-Mn	0.45-0.55	0.60-0.90	0.40-0.60	0.65-0.95
Si-V	0.85-1.00	0.40-0.70	0.50-0.75	0.15 min.

As usual, these are oil quenched and drawn to 388-444 Brinell.

**Stainless Steel Springs.** Tour<sup>1</sup> has studied 0.30 C, 12-13% Cr cutlery stainless from the point of view of springs, by hardness and im-

fact tests, but gives little information on actual spring performance. The Izod bar used was round, 0.40-in. diam., V-notched to 0.25 in. Soaking time before quenching was found to be important; either 1800° for 15 min. or 1725° for 30 min. could be used, but the latter was preferred. The oil-quenched stock is drawn at 500–550° to 45–46 Rockwell C and gives about 4.5 ft.-lb. Tour considers the hardness and impact to be on a par with those of Cr-V springs. If a softer spring, of 37–39 Rockwell C, can be utilized, dropping the C to 0.10–0.12% and drawing at 550–600° gives 10 to 18 Izod.

**Testing for Yield Strength.** In comparing data from different sources it should be noted that the very important property of yield strength is, in steels of this type, not determinable by drop of beam and must be taken from stress-strain diagrams. Authors seldom describe their testing technique in sufficient detail to show just what criterion they used to determine the value given as yield strength or yield point. Thus it is difficult to compare results of different workers. Comparisons are best made on data for several steels from the same laboratory.

**Properties of Some Spring Steels.** According to "Vanadium Steels and Irons,"<sup>2</sup> some V-treated spring steels give the following properties, treated in 0.55-in. diam. Data from Fig. 7, Chap 9, for Al-treated Mo spring steel are added in the last line.

% C	% Si	% Mn	% Cr	% V	Norm. deg.	Quench	Draw deg.	Tensile	Yield	Elong. %	R.A. %	Izod
0.51	0.18	0.68	1.09	0.18	1600–1575	oil	800	225,000	215,000	5	42	11
0.51	0.18	0.68	1.09	0.18	1600–1575	oil	900	200,000	195,000	6	45	12
0.63	0.11	0.72	0.98	0.17	1600–1550	oil	875	225,000	210,000	11	41	10
0.63	0.11	0.72	0.98	0.17	1600–1550	oil	960	200,000	190,000	12	42	14
0.98	0.67	0.49	.....	0.18	1650–1600	oil	950	225,000	200,000	8	20	2.5 <sup>a</sup>
0.98	0.67	0.49	.....	0.18	1650–1600	oil	1025	200,000	180,000	10	23	3.5 <sup>a</sup>
0.70	0.35	0.95	.....	0.10	1550	oil	850	225,000	200,000	11	36	12
0.70	0.35	0.95	.....	0.10	1550	oil	900	200,000	185,000	12	39	15
0.75	0.35	0.75	0.40	0.19	1550	oil	800	225,000	210,000	10	34	11
0.75	0.35	0.75	0.40	0.10	1550	oil	900	200,000	185,000	11	38	16
					Mo							
0.65	0.25	0.80	.....	.....	0.20 1450	oil	800	233,500	215,500	11	41	17
0.65	0.25	0.80	.....	.....	0.20 1450	oil	900	197,000	177,000	11	40	..

<sup>a</sup> As pointed out in Chap. 6, Vol. I, this would be a high quenching temperature for a steel of so high a C content without V. Note the low impact. See also Fig. 3, Chap. 4, for Si-Mn.

**Comparison of Some Spring Steels.** Houdremont and Bennek<sup>3</sup> have listed static properties for a group of spring steels. Comparing these on a basis of 200,000 tensile, on 1/8-in. thick plates, the other properties run as follows:

COMPOSITION					Quench deg.	Temper. deg.	R.A.	
Mn	Cr							
0.90	0.25	0.39	.....	.....	1545 oil	660	172,000	19 31
0.44	1.49	0.60	.....	.....	1525 H <sub>2</sub> O	800	176,000	10 35
0.75	2.70	0.34	.....	.....	1545 H <sub>2</sub> O	975	165,000	15 40
0.61	0.39	1.80	.....	.....	1510 oil	890	177,000	11 30
0.46	0.95	0.92	.....	.....	1545 H <sub>2</sub> O	840	177,000	10 30
0.47	0.23	0.46	1.06	0.16	1650 oil	890	172,000	9 35
0.43	0.17	0.84	0.97	.....	1560 oil	880	170,000	8 32

When 225,000 tensile is sought, the oil quenched low Mn, 0.90% C steel will not meet the requirement at the lowest tempering temperature used. The other steels show:

COMPOSITION					Quench deg.	Temper. deg.	Yield	Elong. ~	R.A.
Mn	Cr								
0.44	1.49	0.60	.....	.....	1525 H <sub>2</sub> O	715	200,000		
0.75	2.70	0.34	.....	.....	1545 H <sub>2</sub> O	860	203,000		26
0.61	0.39	1.80	.....	.....	1510 oil	800	200,000		28
0.46	0.95	0.92	.....	.....	1545 H <sub>2</sub> O	750	196,000		25
0.47	0.23	0.46	1.06	0.16	1650 oil	790	196,000		25
0.43	0.17	0.84	0.97	.....	1560 oil	790	199,000		

Houdremont <sup>4</sup> has also given data for Cr and Cr-Si spring steels:

C	Si	Mn	Cr	Size	Quenched from deg.	Tensile	Yield	R.A.	
0.54	.....	0.84	0.97	0.63 in. thick	1560	780 840	225,000 220,000	200,000 185,000	13 35 14 40
0.63	1.16	0.90	0.57	0.6 x 1.2 in.		930 1000	225,000 200,000	200,000 180,000	10 30 11 32

Hankins and Ford <sup>5</sup> made a study of several spring steels in  $\frac{3}{8}$ -in. thickness by 3 in. wide. An 0.80 C, 0.25 Si, 0.41% Mn steel was included but with this low Mn it could not be quenched to full hardness in oil, even from 1740°, and cracked in water. Hence this was studied at 370 Brinell after oil quenching from 1650°. This hardened the steel

to only about 370, and drawing at 930° did not soften it. A steel of 0.60 C, 0.21 Si, 0.77 Mn with 0.08 Ni and 0.10% Cr did harden and was used to represent C spring steels. The Cr-V and Si-Mn steels are the ones we shall again mention in connection with the work of Hankins and Ford. A low-Cr steel of 0.60 C, 0.26 Si, 0.62 Mn, 0.56% Cr, a high-Cr of 0.45 C, 0.12 Si, 0.69 Mn, and 1.14% Cr, and a Ni-Cr steel of 0.36 C, 0.29 Si, 0.50 Mn, 3.42 Ni, 0.60% Cr were also studied. These steels, oil quenched as shown in the table, were tempered to the same Brinells.

The Si-Mn was also water quenched from 1600°.

Hankins and Ford (and Aitchison had previously stated the same thing) say, "Other things being equal, the steel possessing the highest Izod value appears to be the best one to select for a spring."

**Impact and Endurance of Spring Steels.** On that basis, with special reference to impact, they make the following comparisons:

470 BRINELL			Tensile	Yield	Prop. Limit	Elong. % in 8 in.	R.A. %	Izod	Endurance Limit
	Quench	Draw							
Cr-V.....	1560° oil	890°	235,000	230,000	157,000	5	18	13	98,500
Si-Mn.....	1740° oil	840°	244,000	228,000	152,000	4	21	9	107,500
Low Cr.....	1470° oil	750°	240,000	Not stated	170,000	2.5	2.5	3	105,000
High Cr.....	1510° oil	660°	230,000	230,000	157,000	0.5	1	12	112,000
Ni-Cr.....	1510° oil	660°	220,000	200,000	125,000	4	12	5	112,000
Si-Mn.....	1690° H <sub>2</sub> O	840°	250,000	242,000	186,000	5	27	11	112,000

400-410 BRINELL		Tensile	Yield	Prop. Limit	Elong. % in 8 in.	R.A. %	Izod	Endurance Limit
	Draw							
Cr-V.....	1020°	193,000	186,000	141,000	7.5	33	20	96,000
Si-Mn.....	930°	204,000	188,000	137,000	8	22	12	103,000
Low Cr.....	930°	200,000	Not stated	152,000	5.5	12	9	94,000
High Cr.....	930°	195,000	179,000	134,000	6.5	30	17	92,000
Ni-Cr.....	840°	200,000	190,000	134,000	7	35	12	98,500
Si-Mn(H <sub>2</sub> O).....	930°	204,000	191,000	157,000	8	22	14	103,000

350-360 BRINELL		Tensile Not Stated	Yield Not Stated	Prop. Limit	Elong. % in 8 in. Not Stated	R.A. % Not Stated	Izod	Endurance Limit
	Draw							
Cr-V.....	Over 1110°	(165,000 approx.)	(155,000 approx.)	92,000	(9 approx.)	(35 approx.)	45	92,000
Si-Mn.....	1110°	163,000	146,000	108,000	12.5	30	17	87,500
Low Cr.....	1110°	166,000	Not stated	101,000	6.5	30	22	85,000
High Cr.....	1060°	168,000	157,000	108,000	8.5	35	50	85,000
Ni-Cr.....	1110°	170,000	163,000	121,000	9	33	24	92,000
0.8% C.....	.....	179,000	.....	85,000	9	25	12	87,500
Si-Mn(H <sub>2</sub> O).....	1110°	175,000	155,000	121,000	10.5	27	19	87,500

The endurance limits were determined on polished specimens, not on actual springs. Note that the elongations are in 8 in., not 2 in.

The impact resistance of the high-Cr and Cr-V steels was good at all these hardnesses, somewhat better than Si-Mn, but the endurance of Si-Mn was either better or practically the same as that of Cr-V.

**Fatigue and the Failure of Springs.** Spring failures generally result from fatigue, though the cause of failure in service might well be due to inability to resist occasional overstress rather than to the actual

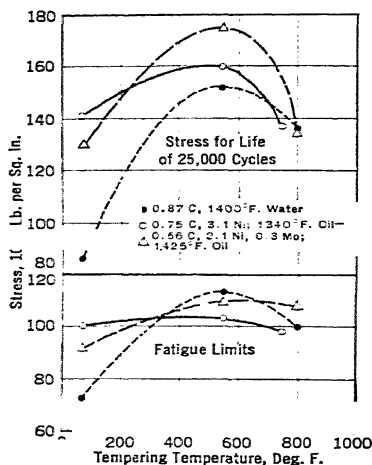


FIG. 1.—Effect of tempering on endurance of martensitized high-C steels. (French.<sup>8</sup>)

position of the endurance limit; and a steel of high impact resistance would be expected to show up well in overstress. However, it is clear that there is no direct relation between impact resistance and the endurance limit itself.

Houdremont and Bennek considered endurance as a basis of evaluation and concluded that (for polished specimens) the composition of the alloy is without effect on the endurance of spring-tempered steel (of equal tensile strengths). This needs to be modified in view of the evidence of Lessells,<sup>6</sup> Gillett and Mack,<sup>7</sup> and French<sup>8</sup> on the effect upon endurance of release of quenching stress by tempering, and French's comparisons of the endurance of water- and oil-quenched, untempered steels of the same hardness. Figure 1 shows the endurance limit and the overstress behavior (life for 25,000 cycles) of three steels. The water-quenched, and hence more highly stressed 0.87% C steel,

at 705 Brinell has much poorer performance than the Ni steel oil quenched to 685 or the Ni-Mo steel oil quenched to 710. On tempering at 550° the C steel has a high endurance limit but low resistance to overstress; but on tempering at about 800°, to produce spring temper, the endurance limits do not vary more than 10,000 lb. per sq. in. and the resistance to overstress is quite similar in all.

The resistance to overstress of the oil-quenched alloy steels drawn at 550° may be ascribed to retained austenite or to smaller quenching stress, with the quenching stress not fully relieved at 550°; but at 800° the austenite should be tempered and the quenching stresses quite fully relieved in all.

**Low Endurance Limits of Commercial Springs.** Hankins and co-workers<sup>a</sup> took up the spring-fatigue problem in detail. They compared Si-Mn and Cr-V spring steel of

C%	Si%	Mn%	Cr%	V%	
0.54	1.95	0.94	....	....	1740° oil
0.55	0.29	0.68	1.16	0.27	1560° oil

Drawn back to 470 Brinell the Si-Mn had 152,000 proportional limit, the Cr-V 157,000, and the endurance limits, on rotating beam specimens polished after heat treatment, were 98,500 and 107,500; but tested as *springs* in rolled, unmachined, heat-treated plates neither one gave safe stress ranges greater than from zero to 45,000.

This led to a study of endurance of the two steels as rotating beam specimens, with the specimens tempered at 930° for the Si-Mn and 1110° for the Cr-V, which brought the Brinells to 450-470 and 380-395, respectively. The quenching temperatures used were respectively much higher for the Cr-V than the 1600-1625° specified by the A. S. M. and somewhat lower than specified for the Si-Mn, and the comparison is marred by the difference in hardness. However, the comparison is of interest.

	Si-Mn	Cr-V
Surface completely polished after treatment.....	103,000	94,000
Normal heating for quenching, not polished.....	60,500	71,500
More heavily decarburized than normal, not polished.....	47,000	45,000
Heated in cyanide, not polished.....	105,000 <sup>a</sup>	98,000 <sup>b</sup>

<sup>a</sup> Brinell hardness surface 460, interior 400.

<sup>b</sup> Brinell hardness surface 550, interior 420.

Repeated bending tests of flat specimens, run to see whether results agreed with those from rotary bending, gave:

	SAFE-STRESS RANGE	
	Si-Mn	Cr-V
Completely polished. . . . .	0-123,000	0-125,000
Normal heating, not polished. . . . .	0- 85,000	0-101,000
Long heating, not polished. . . . .	0- 57,000	0- 83,000
Cyanide heating, not polished. . . . .	0-107,000	0-107,500

**Effect of Decarburization on Endurance of Springs.** The tests show that the difference in the endurance of the two steels is due to differences in surface condition, and that the Si-Mn steel is more prone to decarburize than the Cr-V. If the springs are polished free of the decarburized layer, or if decarburization is prevented, any other differences become very small indeed. The fact is brought out in Chap. 10, Vol. I, that the use of a suitable controlled atmosphere allows heating for hardening without decarburization. That results on spring materials may be extremely variable can be shown by tabulating data from Johnson<sup>10</sup> and from Edgerton<sup>11</sup> for endurance of rod from which coil springs were made, and of the springs themselves. (See p. 311.)

No type of endurance test of the spring stock placed the steels in exactly the same order as did the test on finished springs, which may be affected by differences in decarburization and in grain size.

It is obviously more difficult to secure the necessary yield strength with C steel than with the alloy spring steels, and the latter have superior impact resistance. Among the alloy steels, according to advocates of V,<sup>2</sup> the Cr and Cr-V steels are superior to Si-Mn in that they take less "set," a sort of creep in the cold occurring under repeated stress, by which a coil spring shortens or a leaf spring sags.

Nevertheless there is wide use of Si-Mn or controlled grain size C-Mn springs in the automotive industry; and, if the stock is ground free from decarburization or if decarburization is prevented, most users appear to find that springs of equal yield strength are practically equivalent in ordinary service, if they have had quenching stresses well relieved.

The overwhelming importance of the avoidance of decarburization makes a chapter on spring steels practically a sequel to a discussion of controlled atmospheres for heating for hot working and for quenching without decarburization.

**Possibilities of Austempering Springs.** The fact that there is some suspicion that quenching cracks in martensite needles may not be healed at draw temperatures used for springs raises the question

# ENDURANCE PROPERTIES

311

Spring Steels	C %	Si %	Cr %	V %	Shepard Grain Size	Tensile	ROT. BEAM, ENDURANCE LIMIT		TORSION ENDURANCE LIMIT 0 — MAX		Endurance Limit As Springs	Single Blow Impact <sup>a</sup> Ft.-lb.
							Polished	60° V-Notch	Polished	Notched		
Acid Open Hearth, Carbon.....	0.91	.....	.....	.....	4	218,000	80,000	30,000	102,000	80,000	70,000	18
Basic Electric Carbon.....	1.04	.....	.....	.....	4	237,000	98,000	48,000	123,000	95,000	93,000	10
Cr-V Basic Electric.....	0.52	.....	0.88	0.21	8	237,000	104,000	28,000	128,000	73,000	77,000	15
Si-Mn Basic Electric.....	0.53	1.96	.....	.....	7	236,000	112,000	35,000	138,000	63,000	68,000	18
Basic Open Hearth, Carbon.....	1.05	.....	.....	.....	4	206,000	100,000	64,000	110,000	.....	73,000	..
Electric Carbon.....	1.05	.....	.....	.....	4	205,000	93,000	80,000	90,000	.....	63,000	..
Si-V.....	0.94	0.63	.....	0.17	.....	205,000	102,000	70,000	.....	.....	58,000	..

<sup>a</sup> Round Spec. 0.5-in. diam. 60° V-notch. 0.38-in. diam. at base. 0.03-in. radius at bottom.



whether "austempering" might not give a spring steel of better toughness, substituting primary troostite for secondary troostite, and possibly making austempered C steel, in small springs, as acceptable as alloy steel. Direct evidence is lacking on this point, since recorded experiments are on steels of other C contents and other hardnesses than would normally be used for springs. Davenport, Roff, and Bain<sup>12</sup> hardened and tempered and "austempered" a 0.74 C, 0.37% Mn steel to 50 Rockwell C (see Chap. 5, Vol. I) and materially increased the tensile, yield, reduction of area, and impact in the austempered sample. In spite of the high hardness, the yield strength was not up to spring requirements, though higher Mn or Si might cure this. But at spring hardness of 40–45 Rockwell C the impact of this austempered steel was about 50–55 ft.-lb., while at the same hardness the oil quenched and tempered steel was but slightly inferior in impact, since it gave 40–45. Whether steels that show poor impact at such hardness when quenched and tempered would also fail to show much improvement on austempering remains to be seen. Hughes and Dowdell<sup>13</sup> lead quenched a 0.49 C, 0.78% Mn steel, but the hardness of the 0.505-in. bars used was only about 25 Rockwell C and the results throw no light on the spring problem.

Daasch<sup>14</sup> reports for eutectoid C steel, Mn not stated, tensiles of 170,000–180,000 for hot quenches of 200–600°, with erratic elongation, for 1½-in. bars. Only tensile strength is given. Higher hot-quench temperatures gave lower strength.

Honda and Tamaru<sup>15</sup> used 1½-in. diameter bars of 0.89 C, 0.19% Mn. Quenching into fused salt at 380° C., they developed 174,000 tensile.

Lewis<sup>16</sup> hot quenched 0.192-in. diam. wire of 0.79 C, 0.45 Mn, 0.22 Si, 0.09% Cr. Spring temper was obtained when the salt bath was at 500–600°. He suggested this treatment for springs.

Trautman<sup>17</sup> discusses the heat treatment of 0.60–0.70 C, 0.90% Mn spring wire in 0.091-in. diam. by passing an electric current through it, thus heating it uniformly through its cross section, to 1450°, the operation requiring only 15 sec.; then quenching into a molten, low melting alloy bath held at 425° for about 4 sec., thence to oil where the cooling is completed.

Hardness of 64–66 Rockwell C is obtained. The hardness is reduced to 45–47 Rockwell C by tempering at 800°. Figure 2 compares the critical quenching speed shown by the S curves with the cooling in the alloy and in oil, and indicates that in oil quenching Ar' is not entirely eliminated.

It is shown that the "alloy-quenched" wire will bend upon itself without cracking at 60 Rockwell C.

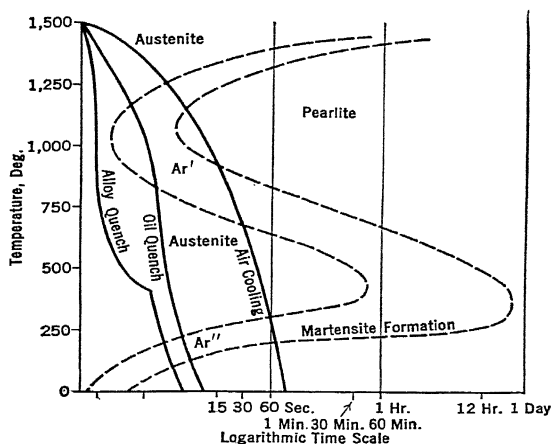


Fig. 2.—Air cooling, oil quenching and molten bath quenching of small diameter spring wire, compared with the "S" curve. Note the logarithmic time scale. (After Trautman.<sup>17</sup>)

The upper dashed curve indicates the beginning of the transformations, the lower one, the end.

Comparative data on 0.091-in. diam. wire of 0.63 C, 1.01 Mn, 0.19% Si, after quenching and tempering to Rockwell C 47–48 by this method and the usual oil-quenching method, are cited as follows:

Quench	Tensile	Johnson Elastic Limit	ELONG. %		R.A. %	Torsion Twists in 10 in.
			10 in.	2 in.		
Oil.....	248,000	203,000	3.4	6.3	53	16
Alloy.....	248,000	229,000	3.1	5.2	55	25

It is stated that owing to the very rapid rate of heating in the electric method, high quenching temperatures, even 1650°, may be used without grain growth. Contrasted with heating the wire by pulling it through a lead bath or through a heated enclosure, the direct electric resistance method heats the whole cross section uniformly instead of the heat having to soak in from the outside. It will be noted that the cross section of the wire treated by alloy quenching was small.

French<sup>8</sup> compared the endurance of a steel of 0.72 C, 2.10 Ni, 0.30% Mo, quenched from 1375° in oil, drawn at 550° for 50 hr. and quenched from 1700° in salt at 580° for 5 min. The oil-quenched and tempered sample had an endurance limit of 110,000, the salt-quenched, one of 90,000, but the hardnesses and static properties are not given. The specimen size is not stated, but was probably around 0.3-in. diam.

On the basis of data so far recorded, austempering does not appear outstandingly advantageous for spring steels, except in very small sections; but there are still possibilities that suitable alloy steels might be chosen for treatment by such a process, which would give improved springs.

The spring steels are "special-purpose" rather than general "constructional" steels. Their automotive use, however, brings them more closely toward the latter category than into the class of the tool and die steels, so they have been discussed in the present section rather than with the other high-C steels of Chap. 21.

**Summary.** Heat-treated springs are made of relatively high-C steel, either plain C or alloy, oil quenched and tempered to secondary troostite with 450 Brinell as an extreme limit of hardness. Since most springs are of relatively small cross section, and the C content is high, the highly alloyed steels of great depth-hardening power are seldom used. Since toughness as high as is compatible with the necessary yield strength (175,000–200,000) is desired, fine-grained steels are favored. For the larger springs that will not harden throughout in C steel, the alloy steels are used, except in some cases like railway springs.

The life of a spring depends on the endurance limit of the finished spring, which is far more dependent on the surface finish and on freedom from decarburization than on the composition of the steel. In heating for forging and for quenching, in the open air or in flue gases, Cr-V steels show less decarburization than Si-Mn or Mo steels. By grinding off the decarburized bark, or by controlled atmosphere heating that avoids decarburization, the controlled grain Si-Mn or Mo steels appear to be practically equivalent, when the surface condition is equivalent, to Cr-V.

The "austempering" process may be applicable for the heat-treatment of springs of small cross section, but what value it may have is limited to quite small sections.

## BIBLIOGRAPHY

1. TOUR, S.: Springs of Stainless Steel, *Iron Age*, v. 138, October 15, 1936, pp. 101-104, 106, 108, 110, 112.
2. Vanadium Corporation of America: "Vanadium Steels and Irons." New York. Author, 1937, 189 pages.
3. HOUDREMONT, E., and H. BENNEK: Federstähle (Spring Steels), *Stahl u. Eisen*, v. 52, 1932, pp. 653-662.
4. HOUDREMONT, E.: "Einführung in die Sonderstahlkunde." Berlin, Springer, 1935, 566 pages.
5. HANKINS, G. A., and G. A. FORD: The Mechanical and Metallurgical Properties of Spring Steels as Revealed by Laboratory Tests, *J. Iron Steel Inst.*, v. 119, 1930, pp. 217-253.
6. LESSELLS, J. M.: Notes on the Fatigue of Metals, *Mech. Eng.*, v. 45, 1923, pp. 695-696.
7. GILLET, H. W., and E. L. MACK: "Molybdenum, Cerium, and Related Alloy Steels." New York, Chemical Catalog Co., 1925, 295 pages.
8. FRENCH, H. J.: Fatigue and the Hardening of Steel, *Trans. Am. Soc. Steel Treat.*, v. 21, 1933, pp. 899-946.
9. HANKINS, G. A., and Co-WORKERS: Research on Spring Steels. See volumes of *J. Iron Steel Inst.*, 1926-1936, inclusive.
10. JOHNSON, J. B.: Fatigue Characteristics of Helical Springs, *Iron Age*, v. 133, March 15, 1934, pp. 12-15; *Ibid.*, March 22, pp. 24-26.
11. EDGERTON, C. T.: Abstract of Progress Report No. 3 on Heavy Helical Springs, *Trans. Am. Soc. Mech. Eng.*, v. 59, 1937, pp. 609-616.
12. DAVENPORT, E. S., E. L. ROFF, and E. C. BAIN: Microscopic Cracks in Hardened Steel. Their Effects and Elimination, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 289-310.
13. HUGHES, T. P., and R. L. DOWDELL: Quenching Steel in Hot Lead, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 737-750.
14. DAASCH, H. L.: Quenching Steel in Hot Baths, *Metal Progress*, v. 24, November, 1933, pp. 27-29.
15. HONDA, K., and K. TAMARU: On a New Method of Quenching Steels in a High Temperature Bath, *Trans. Am. Soc. Steel Treat.*, v. 13, 1928, pp. 95-104, 125.
16. LEWIS, D.: The Transformation of Austenite into Martensite in a 0.8 Per Cent Carbon Steel, *J. Iron Steel Inst.*, v. 119, 1929, pp. 427-441.
17. TRAUTMAN, O. C.: Electric Heat Treating of Wire, *Iron Age*, v. 138, October 15, 1936, pp. 34-39.

## CHAPTER 17

### STEEL CASTINGS

**Similarity of Alloying Effects in Wrought and Cast Steels.** There is no sharp dividing line between wrought and cast steels in relation to heat treatment. The strengthening and toughening tendencies of the different alloying elements are exerted in the same directions in both; hence the cast steels have been discussed along with the wrought steels in most of the previous chapters dealing with each element and combinations of elements. Only a brief recapitulation will be given here.

**Differences in Suitability for Quenching.** The chief difference between wrought and cast steels lies in the fact that the hot working of wrought steels breaks up the dendritic ingot structure, renders the steels more readily homogenized on heating into the austenitic range, and gives a mechanical refinement of the grain that adds to ductility and toughness. The limiting C content of cast steel is lower than that of forgings from comparable steels because ductility must be preserved. The irregular section of many castings renders them difficult to quench without cracking, so that martensitic hardening followed by tempering is less readily applied to castings as a class than to wrought objects; although, as Lorenz<sup>1</sup> points out, quenching and tempering are being increasingly and very successfully applied nowadays in cases where not long ago such treatment would have been thought impractical. However, castings are often so large that no sufficiently cheap moderately alloyed steel can be found that will harden satisfactorily in such sizes, the properties obtained by quenching and tempering being little better than those obtained by normalizing and drawing.

**Normalizing, the Standard Treatment for Cast Steel.** In search of ductility, it was formerly the practice to anneal steel castings, but this practice has been generally replaced by normalizing, since normalizing of most cast steels gives ductility and toughness quite on a par with annealing without as much sacrifice of yield strength. There are, of course, cases where full annealing or the use of a spheroidizing anneal are required. On the average, however, a single or double normalizing followed by a draw gives the best combination of properties when quenching and tempering are not applicable. The selection of alloy-

ing elements for cast steel might therefore be phrased as the problem of alloying for normalizing.

**The Cost Factor.** Raw material costs form a larger proportion of the total cost of steel castings than is normally the case with objects made of wrought steel. Expensive alloying elements are, therefore, dispensed with as far as possible. The cast steels are usually made in relatively small quantities, compared to tonnage open-hearth wrought steel, and precise analytical control of composition is more difficult and more costly because of the quicker heats and the smaller tonnage per analysis. Alloy steels in which some variation in composition can be allowed without marked change in properties are preferred. To secure sound castings, very large heads and risers must be used, leaving large amounts of metal to be remelted; hence expensive elements that are readily oxidizable, and so would be lost in the remelt, are avoided when feasible. The Cr, Si, Mn, and V contents are lowered on remelting under oxidizing conditions. As will be brought out in the next chapter, of the strengthening elements, Mn, Si, Cu, and Cr are the cheapest and Ni the most costly. Of the grain-refining elements Al is the cheapest, Ti and Zr next, Mo next, and V the most costly; but the amounts required are small. The tendency is to depend on Mn for primary alloying effect, even in "plain" cast steel.

The normal Mn and Si contents are higher than in wrought "plain" steel; 0.30–0.50 Si and 0.60–0.80% Mn are generally used in order to secure sound castings and such amounts have definite strengthening effects.

**Manganese.** To produce ductile steel of higher strength than can be obtained without alloying, the most common procedure is to raise the Mn to somewhere between 1.00 and 1.50%, and to supplement the Mn with either a ferrite-strengthening element like Ni, Cu, or Si, or a grain-refining element like Mo, V, or Ti, or both. As Mn and other alloying elements are increased, the C, which may run up to about 0.35% in plain C steel, is generally reduced nearer to 0.25% to maintain ductility. Since Mn is the leading alloying element, in cast steels, the properties of some of the most important commercial types of alloy cast steels have been discussed in Chap. 3.

**Silicon.** Silicon is not often raised much above 0.50%, though excellent properties may be had in steels with 1.00–1.50% Si. The addition of Si raises the critical temperature and the necessary temperature for proper normalizing by 200–300°. Since even C steel castings have to be heated to a high temperature for proper homogenization of the cast structure, the straight Si steels require temperatures that make the castings prone to scaling and warping during heating.

Nevertheless it will be recalled that high Si serves excellently in Finlayson's Cu-Mn-Si steels (Chap. 5).

**Chromium.** Chromium, along with high C, finds some use in castings for wear resistance when extreme toughness is not essential; 0.45 C, 0.75% Cr is a common composition, but because of the loss of toughness that accompanies the gain in strength due to Cr, the plain Cr steel is not a common one for castings. When castings can be quenched and tempered, Cr for depth hardening, plus Ni for toughness, make a desirable combination. For normalizing or annealing, however, some of the lower cost steels give even better properties. Armstrong met trouble with 0.30 C, 1.50 Ni, 0.60% Cr steel cracking in the mold.

**Copper.** Copper cast steels are relatively new, but, as is brought out in Chap. 5, they have very interesting possibilities when the Cu is accompanied by small amounts of other alloying elements.

**Nickel.** The plain Ni cast steels, of course, have good toughness and are widely used in castings for severe service. Because of cost, the 3.50% Ni cast steel is quite generally replaced by steels with part of the Ni substituted by Mn, as discussed in Chap. 3. Nickel-V cast steel is one of the most reliable compositions, and would have wider use save for the cost factor which is the prime reason for the substitution by Mn. Partial substitution by Cu would also be feasible.

**Molybdenum.** Molybdenum in cast steel is ordinarily added along with other alloying elements. It confers strength (especially at high temperatures), it corrects for temper brittleness after the draw in normalized and drawn Mn, Ni, or Cr steels, and it gives depth hardening in steels for quenching. It also gives some grain-size control, allowing high-temperature normalizing without excessive coarsening. A grain-control addition is generally needed in normalized alloy cast steel.

**Grain-control Additions.** Vanadium has been the mainstay for grain-size control; but its cost, plus the rapid coarsening of V steel once the coarsening temperature has been exceeded, have caused the foundryman to study Ti, which seems especially effective. Zirconium has a rather similar action.

Attention is now being paid to the study of multiple additions, i.e., of more than one of the grain-controlling additions. Aluminum, as has been mentioned in Chap. 8, is added to most cast steel, in very small amounts, with the sole object of avoiding blow holes. There are possibilities in larger additions for grain control.

**Welding of Castings.** There is a growing tendency to build up structures by welding steel castings and rolled plates or shapes where such construction is more economical than making the whole structure

as a large casting; hence, the alloy additions may have to be considered from the standpoint of the suitability of the castings for welding. Spraragen and Claussen<sup>3</sup> have summarized information on this subject.

**The Usual Alloying Principles Hold for Cast Steels.** The alloying principles are essentially the same as those for wrought steels. Strengthening of the ferrite by a suitable alloy; stabilization of the carbide by the same alloy (in the case of Mn) or by another; facilitating the preservation of fine grain, even on heating the austenite to a high temperature, by some addition that raises the coarsening temperature; and the necessity of lowering C when the alloy content is built up—all are recognized principles in wrought steel.

Lack of working; larger sections, on the average, than are common in parts made of wrought steel; and the usual necessity for confining the heat treatment to normalizing, these tend to modify slightly rather than to change radically the conclusions drawn from experience with wrought steels. Since the heating for, and cooling in, normalizing are the processes that primarily control quality in cast steels, the precautions mentioned in Chaps. 16 and 17 of Vol. I are important.

For a comprehensive summary of the properties of, and treatments for, alloy cast steels, the articles by Lorig and Williams, Grotts, Strauss, Hall and Lorenz in the 1932 A. F. A.-A. S. T. M. "Symposium on Steel Castings"<sup>4</sup> should be consulted. Essentials that must be met in the treatment of both C and alloy steel castings have been mentioned in Chap. 1.

**Flexibility.** The steel foundryman is not severely limited in his choice of compositions. Cast steel is made in small heats, and the purpose for which the casting will be used is always known. If a non-standard composition seems called for, for a specific purpose, it can be produced and applied much more readily than in the case of wrought steels where the steel maker and the jobber stock rolled bars or forging billets only of the accepted tonnage compositions for which many orders are expected. Making a small heat of special steel and keeping it separate during rolling involve extra trouble and expense. The forge shop and the heat treater are seldom in a position to order a small lot of special analysis unless the need for it is overwhelming. The steel foundry, on the contrary, can make its own special steels in convenient quantities and can more readily develop "tailor-made" steels for individual applications.

**Composition Is Not All in Casting.** It must be remembered that neither the alloy nor the heat treatment entirely governs the properties of the finished casting. Melting with such practice that porosity is



avoided, gating and risering so that shrinkage cavities or checks are not present, and, above all, design of the part to avoid having to feed a thick portion through a thin one, which makes shrinkage defects inevitable, must be attended to. Failure to avoid these difficulties cannot be compensated for by use of alloy steel nor trick heat treatment. The American Foundrymen's Association has recently published a 62-page pamphlet "The Influence of Design on the Stress Resistance of Steel Castings," prepared by R. A. Bull, which deals with these topics and expands the previous article by Bull in the A. F. A.-A. S. T. M. Symposium.<sup>4</sup> See also Briggs.<sup>5</sup>

#### BIBLIOGRAPHY

1. LORENZ, F. A., JR.: Notes on the Design of Steel Castings, A. F. A.-A. S. T. M., "Symposium on Steel Castings." 1932. pp. 20-38.
2. ARMSTRONG, T. N.: Properties of Some Cast Alloy Steels, *Trans. Am. Soc. Metals*, v. 23, 1935. pp. 286-318.
3. SPRARAGEN, W., and G. E. CLAUSSEN: Welding Cast Steel—A Review of the Literature to November 1, 1936, *Welding J.* (New York), v. 16, April, 1937, pp. 2-11, Suppl.
4. American Foundrymen's Association and the American Society for Testing Materials: "Symposium on Steel Castings." New York, The Societies, 1932. 254 pages.
5. BRIGGS, C. W.: Steel Casting Design for the Engineer and Foundryman, *J. Am. Soc. Naval Eng.*, v. 50, 1938, pp. 173-230.

## CHAPTER 18

### FACTORS IN THE SELECTION OF CONSTRUCTIONAL STEELS

Before passing on to the special-purpose tool and die steels and other special, highly alloyed steels, which are primarily judged on other criteria than the static tensile properties, we may pause to consider briefly the problem of a choice among the constructional alloy steels.

**The Cost Factor.** If all the metallic elements cost the same amount per pound or per unit volume, the engineer could select his alloy steels solely on the basis of what the alloying elements accomplish. Since they do vary widely in cost, he must figure on what the steels accomplish per dollar.

**Raw Material Cost.** The raw material costs per pound of contained metal may be compared as follows on the basis of prices \* at the time of writing:

Fe—As soft C steel bars, not alloy grade, base price.....	\$0.0225-0.0235
Alloy grade.....	0.028
Mn—As high-C, 80% ferro.....	0.051
Si—As 50% ferro.....	0.062
Cr—As high-C ferro.....	0.105
Cr—As low-C (0.06%) ferro.....	0.20 <sup>a</sup>
Cu—As virgin metal.....	0.11½ <sup>b</sup>
P—As ferro.....	0.145
Al—As No. 2 metal.....	0.14 <sup>c</sup>
Ni—As virgin metal.....	0.35 <sup>a,d</sup>
Ti—As high-C ferro.....	0.40
Zr—As ferro.....	0.40
Mo—As molybdate.....	0.80
Mo—As ferro.....	0.95
Co—As metal.....	1.36 <sup>e</sup>
W—As ferro.....	1.60-1.65 <sup>f</sup>
Cb—As ferro.....	2.25
V—As ferro.....	2.70

<sup>a</sup> Lippert <sup>1</sup> states the 1937 price of Ni as 30 to 32 cents per lb., depending on quantity, and of low-C ferrochromium as 19.5 cents.

<sup>b</sup> Fluctuations from \$0.05 to \$0.17 have been met in recent years. Scrap is often obtainable a little under the new metal price.

<sup>c</sup> Secondary metallurgical ingot or new clippings may be obtainable at slightly lower figure.

<sup>d</sup> Scrap prices slightly lower, but scrap not widely available. Nickel along with Cu, which can often be utilized as replacement for part of the Ni, could be added as Monel metal of about 68 Ni, 29 Cu, 1.5 Fe, 1.1 Mn, 0.1 Si, 0.15% C. New metal price \$0.28, or approximately the same as the cost of the contained Ni and Cu. Monel turnings are sometimes available at a considerable reduction.

<sup>e</sup> Comstock, G. F., *Metals and Alloys*, Oct., 1938, p. 287 puts the price of commercial Co at \$1.75.

<sup>f</sup> Present price up from that of about \$1.30, two years ago.

\* *Iron Age, and Daily Metal Trade*, September 15, 1938. Cu price as of October 19, 1938.

## 322 FACTORS IN THE SELECTION OF CONSTRUCTIONAL STEELS

**Alloy Extras.** "Alloy extras" are added to the base price of steel to give the cost of an alloy steel in bar form. These spring from the

## ALLOY CONTENT EXTRAS, HOT ROLLED

S.A.E. AND SPECIAL GRADES (Chemical ranges expressed in per cent)		Open- hearth Bars—Extra Per 100 lb.	Electric- furnace Bars—Extra Per 100 lb. <sup>a</sup>
T-1300 (Mn 1.51-2.00).....	1.75 Mn.....	\$0.10	
T-1300 (Mn 1.51-2.00 C 0.20 max.)	1.75 Mn low C.....	0.35	
2000.....	0.50 Ni.....	0.35	
2100.....	1.50 Ni.....	0.75	
2300.....	3.50 Ni.....	1.55	
2500.....	5 Ni.....	2.25	
3100.....	1.25 Ni, 0.60 Cr.....	0.70	
3100 (Ni 1.25-1.75).....	1.50 Ni, 0.60 Cr.....	0.80	
3200.....	1.75 Ni, 1.10 Cr.....	1.35	
3300.....	3.50 Ni, 1.50 Cr.....	3.80	
3400.....	3 Ni, 0.75 Cr.....	3.20	
4100 (Mo 0.15-0.25).....	1 Cr, 0.20 Mo.....	0.55	
4100 (Mo 0.25-0.40).....	1 Cr, 0.35 Mo.....	0.75	
4345.....	1.75 Ni, 0.75 Cr, 0.25 Mo.....	1.65	
4340.....	1.75 Ni, 0.65 Cr, 0.35 Mo.....	1.85	
4600 (Ni 1.50-2.00, Mo 0.20-0.30).....		1.10	+0.35
4800.....	3.50 Ni, 0.25 Cr.....	2.00	+0.40
Low-Ni-Mo (C 0.50-0.90, Cr 0.15 max., Ni 0.55-0.80, Mo 0.10-0.25)		1.00	
5100 (Under 0.60 Cr).....		0.25	
5100 (Cr 0.60-0.90).....		0.35	
5100 (Cr 0.80-1.10).....		0.45	
5100 (Spring Flats).....		0.15	
6100 (Bars).....	1 Cr, 0.16 V.....	1.20	
6100 (Spring Flats).....	1 Cr, 0.16 V.....	0.85	
9200 (Spring Flats).....	2 Si.....	0.15	
9200 (Spring Rounds and Squares).....	2 Si.....	0.40	
9255 (Common Tool Grade).....		1.25	+0.75
C-Mo (Mo 0.10-0.20).....		0.35	+0.40
C-Mo (Mo 0.15-0.25).....		0.40	+0.40
C-Mo Spring Flats (Mo 0.10-0.25).....		0.15	<sup>x</sup>
C-Mo Common Tool Grade (C 0.65-0.75, Mo 0.15-0.25)		0.90	+0.40
C-V (0.15 V).....		0.85	
Ni-Cr-V.....		1.50	
2½% Ni-Mn (R. R. Analysis).....		1.20	
Silchrome (C 0.50 max., Si 3.60-4.20, Cr 1.85-2.50).....		2.60	+0.25

<sup>a</sup> The extra for electric furnace bars is \$0.50 per 100 lb. added to that for open hearth, as noted.

<sup>x</sup> No electric furnace extra quoted, steels normally made in the open hearth.

increased cost due to the alloying elements as well as to the added control, analysis and testing, the smaller recovery of sound metal, the loss of alloy in remelting scrap, the extra power involved in rolling, and other similar factors. The base price for hot-rolled, open-hearth alloy steel bars is \$0.028 per lb. The schedule of extras on p. 322 for alloy content has been in effect in recent months, on constructional alloy steels.

**Prices of Stainless and Rustless Bars.** The base prices \* for hot-rolled bars of the highly alloyed, corrosion-resistant rustless and stainless steels are given below:

No.	C	Cr	Ni	Base Price, Cents per Pound
410	0.07-0.12	12.00-14.00	.....	18.5
430	0.12 max.	16.00-18.00	.....	19
442	0.35 max.	18.00-23.00	.....	22.5
446	0.35 max.	23.00-30.00	.....	27.5
302	0.20 max.	17.00-19.00	7.00-9.50	24
304	0.08 max.	17.00-19.00	7.00-9.50	25

The comparison of prices may be put on another basis, that of "warehouse base prices for steel delivered locally." Such prices for some of the S.A.E. steels, as unannealed, hot-rolled bars, in Detroit, are,† in cents per pound:

S.A.E.	1035-1050	3.83
	2300 series	7.57
	3100 "	6.17
	4100 "	5.92
	6100 "	7.39

**Relation of Heat-Treating Methods to Choice of Steel.** The final properties of the heat-treated part are not the only criteria in the choice of the steel for the part. The steel must not only fit that need, but it must be amenable to the methods of forming and of heat treatment that one desires to use.

Carburizing steels are often chosen, not on the sole basis of case hardness and core toughness after a complex series of heat treatments designed to bring out the best properties as measured by tensile, Rockwell, and Izod tests, but to get a usable combination of those properties

\* *Steel*, September 19, 1938.

† *Steel*, September 19, 1938.

with minimum warpage. (Minimum warpage is obtained on quenching direct from carburizing without reheating, or at least by one of the simple treatments.) So the real choice in such a case might be confined to the steels suitable for a single quench.

Spring steels, as stated in the previous chapter, are perhaps to be evaluated on propensity toward decarburization as much as on any other one factor, if the surface is not to be ground. Both Boegehold<sup>2</sup> and Peterson<sup>3</sup> indicate that the high Si-Mn spring steel is likely to be chosen when the surface is ground so that this disturbing factor is eliminated. From such comments as theirs and the comment of Graves<sup>4</sup> it becomes evident that the large users are tending to employ constructional steels alloyed with Mn instead of Ni, and those made fine-grained with Al instead of with V, for price reasons, although in less precisely controlled heat-treating operations a small user might find it to his advantage to use the more expensive steel. Yet, when a given set of properties is needed that cannot be got from a C or a cheap alloy steel, the large users do not hesitate to specify steels with Ni, V, or Mo, in spite of the added cost, as the tonnage of such steels testifies. They do take care, however, not to use a higher alloy content than is needed and to adjust the C content in relation to the alloy content so as not to sacrifice toughness while building up the strength.

Hall<sup>5</sup> remarks, in connection with chisels, that if the plant has a blacksmith who judges temperature by his eyes, and his eyes happen to be getting aged, it is better to give him a Cr-W-V chisel steel; but, if the plant has pyrometers and the hardener is very good, it is fairly safe to sell them a Si-Mn chisel steel at half the price, even though it has a very narrow hardening range. This same principle holds throughout the whole range of choices.

When each heat of steel comes into the user's plant separately marked so that it can be checked through a series of tests, and the normalizing practice, time, and temperature of heating for quenching, and the tempering time and temperature are adjusted to meet the particular composition and grain-coarsening characteristics, the desired results will be far more certainly obtained than when the steel is merely given a treatment recommended by the steel maker, or in some handbook, for that general grade of steel. Even though the hardener may not be able to apply exhaustive tests, a few specimens hardened from different temperatures (Rockwell, Vickers, or file tested for hardness, and examined for the nature of the fracture, then tempered, tested for hardness and for toughness, even though the equipment for the last consists only of a vise and a hammer) will allow him to approximate the properties wanted more closely than is probable by blind adher-

ence to a set of directions. Steel of recorded composition, hardenability, and grain-coarsening characteristics is now obtainable, even in small lots, from certain warehouses.

The element of personal judgment bulks large in fitting the steel and its heat treatment to the particular need, according not only to the engineering requirements of the piece itself but also to the available heat-treating facilities and with due regard to the specific requirement for machinability, freedom from distortion, soft skin, etc. At the same time the initial cost should be balanced against the fabrication cost so that one may not be penny wise and pound foolish.

**Standard and Special Compositions.** The S.A.E. steels cover the main requirements, but individual needs may dictate a smaller C range, a different ratio of alloying elements, or the development of a steel not on the list. Some steels not on the S.A.E. list are in as great demand and as available as some of those on the list. These other steels were developed because a steel maker or a user fitted his knowledge of the properties needed, and the ease of obtaining them by economical practice, to the alloying potentialities of the different elements. At first tailor-made for one specific need, such steels in time became the ready-made steels, available for analogous uses.

**Shifting Conditions Shift the Choices.** Improvements in steel-mill technique, in heat-treating furnaces, in machining practice, and so on, constantly shift the suitability of different steels to different uses so that the choice of yester-year may not be wisest now. Any dogmatic statement of the superiority of one steel over a competing one is likely soon to become untrue, even though it may have been accurate under the previous conditions.

There is no likelihood that the choice of constructional steels will ever become so fixed that the exercise of engineering and metallurgical judgment can be dispensed with. In exercising this judgment the choice is none too easy when only the properties that are commonly specified are considered, since, especially in the easily machinable range, practically equivalent results are given, in these common tests, by a large number of steels.

**Yield Strength and Modulus of Elasticity.** If mere stiffness is required, since the modulus of elasticity is scarcely affected by alloying or by heat treatment, the only requirement is that the yield strength be high enough to prevent plastic distortion within the loading range. Up to the yield strength, one steel is as stiff as another. If ductility played no rôle in service, one could compare steels solely on the basis of yield strength. However, on cursory examination, even the inclu-

sion of ductility requirements still leaves many steels on a par with each other.

**Replacement of One Alloy Steel by Another.** To pick an example, according to Dawe <sup>6</sup> 5140, plain Cr, 4140, Cr-Mo, and 6140, Cr-V, give, after quenching and drawing at three temperatures:

	5140	0.40	0.90						
	4140	0.40	0.90	0.20% C	Mo	Oil quenched			
	6140	0.40	0.90	0.15% C	V				
	Draw 800°			Draw 1000°			Draw 1300°		
	5140	4140	6140	5140	4140	6140	5140	4140	6140
Tensile	187,000	210,000	207,000	147,000	164,000	162,000	99,000	121,000	108,000
Yield	168,000	180,000	185,000	125,000	145,000	142,000	78,000	101,000	98,000
Elong. %	12	12	13	16	15	15	23	25	23
R.A. %	44	47	46	53	55	55	65	68	66

Such comparisons could be extended to cover the whole gamut of alloy steels, and data for a wider range have been assembled.<sup>7, 8</sup> One can thus show that, within the range of sizes that will harden completely, the static tensile properties of quenched and tempered steels, tempered to the same tensile strength, may not differ enough to offer engineering justification for a choice of one over another on the basis of tensile properties alone. Clapp <sup>9</sup> has made this argument, collecting, as in the table, p. 327, the yield strengths and elongations for several S.A.E. steels. Superficially, the argument seems sound, but on further examination it will be found that at least minor differences do exist. It may be noted that reduction of area is omitted in Clapp's comparison.

**Ductility.** Several high authorities have examined and discussed the fine points of differentiation.

Grossmann <sup>10</sup> provides curves, after Janitsky, from which one can assemble as in Fig. 1 the tensile properties for 0.30% C water-hardening steels 1330, 2330, 3130, 4130, and 6130, and 0.40-0.45% C oil-hardening steels 2345, 3145, 3240, 4145, 4340, 4645, and 6145, on the basis of tempering to equal tensile strengths, for 1-in. round specimens. In the region where the steels are fairly readily machinable with ordinary high-speed tools it is very difficult to make any choice simply on the basis of tensile properties, in small sizes.

However, the oil-quenched steels and those drawn at the highest temperatures tend to be better in reduction of area at tensile strengths above 175,000. Boegehold <sup>2</sup> emphasizes this in Fig. 2.

CLAPP'S TABLE FOR YIELD STRENGTH AND ELONGATION OF S.A.E. STEELS  
DRAWN TO THE SAME TENSILE STRENGTH

Tensile	1040	2340	T1340	3140	6140	5140	3240	4640	4340	4140
110,000.....	78,500 18.5	90,000 24.5	92,000 28	92,500 24.5	92,000 24					
130,000.....	.....	110,000 22.5	110,500 20.5	111,000 21	114,000 20.5	107,000 21.5	109,000 22	114,000 23.5	103,000 21.5	108,000 21
150,000.....	.....	130,000 20	127,500 15.5	130,000 17.5	135,000 17	127,500 18	129,000 19.5	135,000 20	130,000 19	128,500 15.5
170,000.....	.....	149,000 16	144,000 12.5	148,000 15.5	154,000 15	148,000 16	149,000 17.5	155,000 15	152,000 17	152,000 13
190,000.....	.....	.....	.....	.....	173,000 13.5	169,000 12.5	170,000 16.5	175,000 13	172,000 15	173,500 12
210,000.....	.....	.....	.....	.....	.....	.....	.....	.....	192,000 12	191,500 12
Quench Temp. Deg.	1550	1475	1500	1500	1600	1550	1500	1500	1550	1575



**Impact.** Even greater differences appear when steels of equal tensile strength are compared on the basis of impact values. Figure 3,

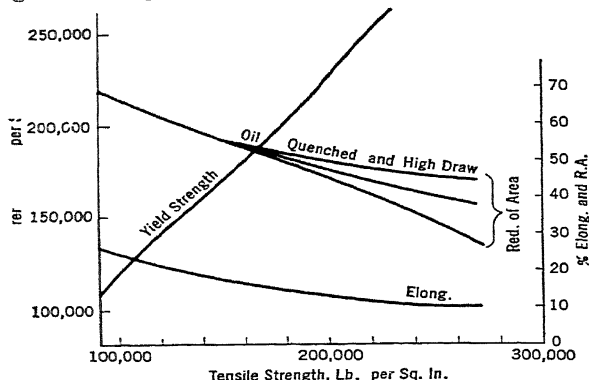


FIG. 1.—Comparison of quenched and tempered alloy steels, 0.30% C steels water quenched, 0.40–0.45% C steels oil quenched, drawn to a given tensile. (Grossmann,<sup>20</sup> after Janitsky.)

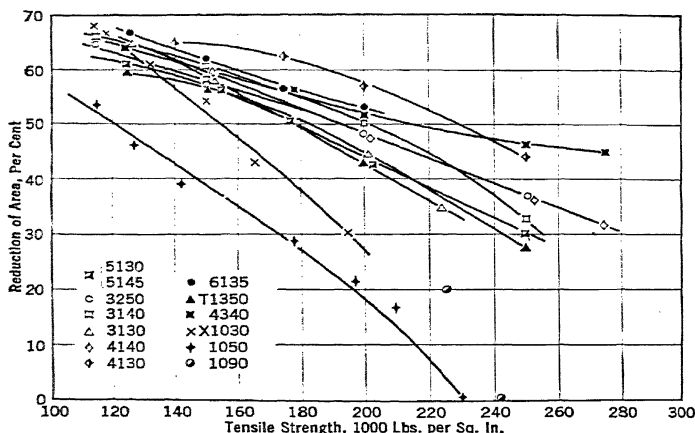
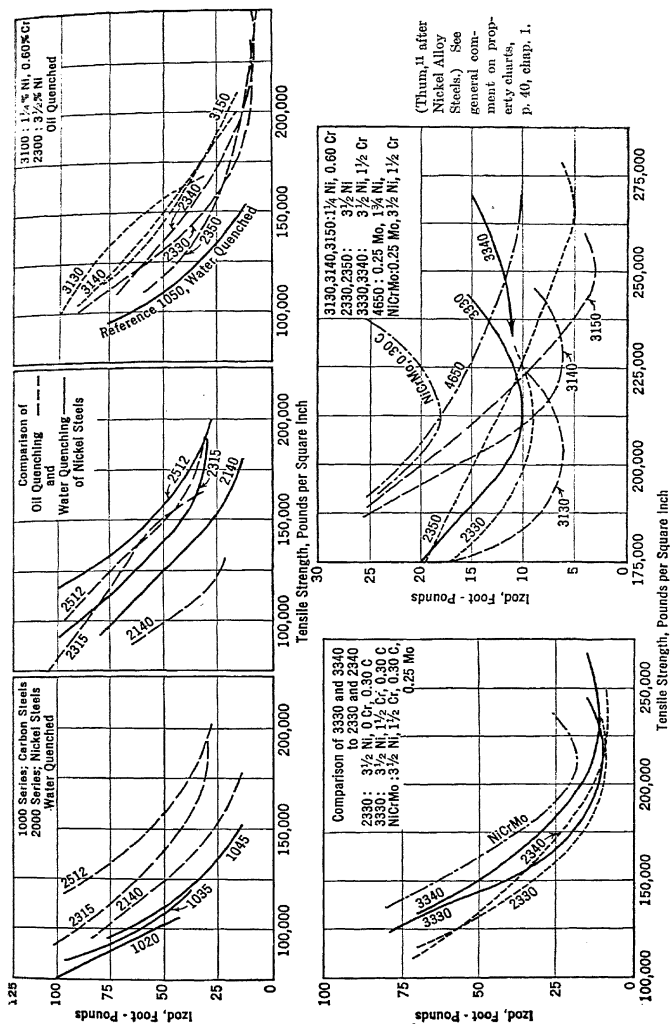


FIG. 2.—Comparison of R. A. for steels heat treated to the same strength and hardness. (Boegehold.<sup>21</sup>)

Specimens treated as 1-in. rounds; except X1030, 1050 and 1090 treated as 0.51 in., and 4130, treated as 1.33-in. rounds. All oil quenched except 4130 and X1030, which were water quenched.

The alloy steels show greater ability to neck down than do the C steels.

assembled by Thum<sup>11</sup> from data of French and Sands, brings this out for Ni steels. Strauss<sup>12</sup> has assembled data from which Fig. 4 is



(Thum, 11 after  
Nickel Alloy  
Steels.) See  
general com-  
ment on prop-  
erty charts,  
p. 40, chap. 1.

FIG. 3.—Impact resistance of quenched and tempered C and alloy steels at various tensile strengths.

# 330 FACTORS IN THE SELECTION OF CONSTRUCTION STEELS

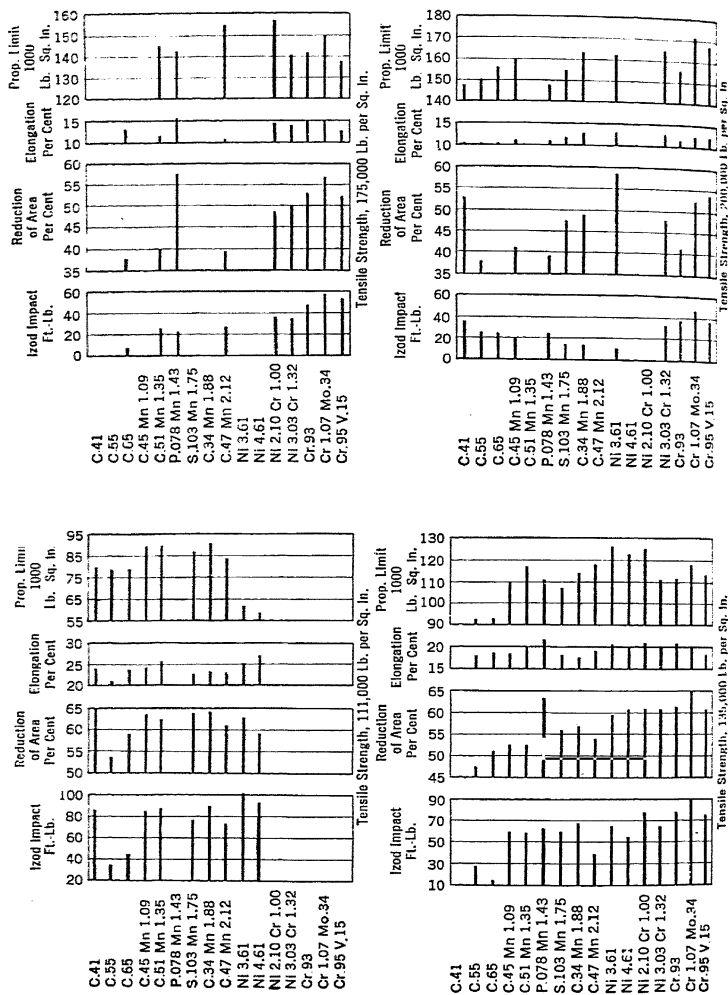


Fig. 1 Properties of different steels heat treated to the same tensile strength.

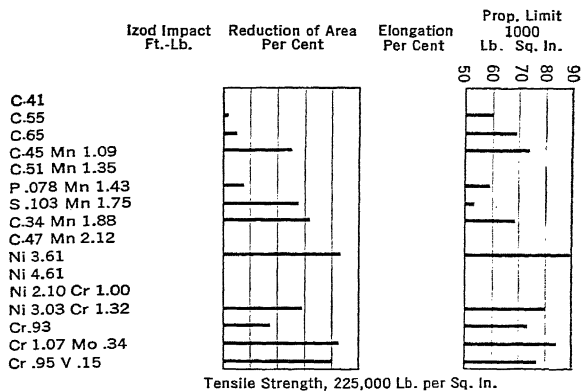


FIG. 4.—Continued.

Composition:						Composition:						
C	Mn	Ni	Cr	Mo	V	P	C	Mn	Ni	Cr	Mo	V
0.41	0.67	....	....	....	....	0.10	0.47	2.12	....	....	....	....
0.55	0.36	....	....	....	....	0.05	0.27	0.92	3.61	....	....	....
0.65	0.65	....	....	....	....	0.05	0.23	0.61	4.61	....	....	....
0.45	1.09	0.18	....	....	....	0.05	0.44	0.43	2.10	1.00	....	....
0.51	1.35	....	....	....	....	0.05	0.42	0.25	3.03	1.32	....	....
0.35	1.23	0.32	....	....	....	0.05	0.55	0.76	....	0.93	....	....
0.38	1.75	....	....	....	....	0.05	0.37	0.65	....	1.07	0.34	....
0.34	1.88	....	....	....	....	0.05	0.30	0.70	....	0.95	....	0.15

See general comment on property charts, p. 40, chap. I.

drawn, for steels of equal tensile strengths which show decided differences in impact, even though the elongation and yield strength do not vary much. Thus, as we examine the properties more closely, greater differences appear than are brought out in a superficial comparison like Clapp's table.

**The Size Effect.** In larger sections it is imperative to use the more highly alloyed steels if we wish to get high strength at the center of the section. This has been made evident in the discussion of critical cooling rate and depth hardening and represents the primary reason for the introduction of alloying elements. Still, in any size, there is some choice among several steels, each of which can be made, by proper selection of quenching temperatures, quenching medium, and tempering temperature, to give equivalent tensile values.

**Subtle Differences.** The actual choice is really made on the basis of more subtle differences which may relate either to the performance of the part in service, or to convenience somewhere along the line in fabrication.

**Uniformity and Cleanliness.** The uniformity with which the steel maker can produce successive heats of similar grain coarsening and

hardenability properties varies somewhat with the type of steel. The number and type of non-metallic inclusions in the steel are affected by the alloying elements present. One steel may be more prone to seaminess or to cracking in rolling than another. Such factors affect the cost of steel to the purchaser, as thoroughly as does the mere cost of the alloying elements themselves; and, if these and allied factors, in spite of the watchfulness of the steel maker, cause faulty material to get to the user, the user is affected. Johnson<sup>13</sup> compares the endurance limits of four steels whose compositions fall within the limits 0.40-0.46 C, 0.73-0.86 Cr, 1.47-1.96 Ni, heat treated to give 148,000-152,000 tensile, 135,000-140,000 yield, 17-19% elong., 300-325 Brinell, 45-60 Izod. The amount and type of inclusions varied. The endurance limits varied accordingly, as follows:

Inclusions	Endurance Limit
Very few and very small. ....	80,000
More, but small and rounded. ....	79,000
Still more, strung out. ....	69,000
Still more, larger, elongated. ....	64,000

Such evidence leads Johnson to conclude that "a low inclusion count is an additional safeguard in the selection of steel." It is indicated, however, that actual fatigue failure far more often arises from laps, pits, external notches, and poor fillets, than from inclusions.

**Machinability.** Some of the subtle differences are hard to measure. Under a given set of machining conditions a user may find that the shop insists that one steel acts better than another, although the actual proof of this may be difficult. Graves<sup>4</sup> comments that in drilling piston pins made from one instead of another of the more machinable of the steels suitable for such use, a saving in man time of 8 sec. per car is made, while the extra cost of the steel is 1 cent per car. Telling whether the saving is worth the cost involves hair-splitting accounting methods.

**Automotive Choices.** Graves<sup>4</sup> discussed the steels used for various parts in half a dozen makes of passenger cars, and Boegehold<sup>2</sup> and Peterson<sup>3</sup> have given somewhat similar discussions. Steering knuckles, quenched and tempered, are made from T1330 or T1340, 4130, 3140, 5130, or C-Mo, all at 240-300 Brinell, machined after heat treatment. The strength and toughness are alike and the choice is made on what the different plants believe as to machinability; obviously their beliefs differ. Steering knuckle pins, all carburized, are made from X1015, X1314, 3115, 4620, 6120, or C-Mo. Steering worms are made from 5130 or 5135 cyanide cased or from 4615 carburized. That only

two steels are used is explained by Graves on the basis that each is used by large producers of steering gears. Automobile makers have followed their lead, not wishing to take any chances.

Rear axle shafts are made from 1040, T1330, 4140, or C-Mo, and Graves says that as the lower alloy and plain C steel are improved in *uniformity* their use will increase.

**Stress Concentration and Fatigue.** Knowlton<sup>14</sup> has discussed the choice of steels for splined tractor axle shafts in relation to variations in hardness and toughness and to shaft design. He made extensive studies on 3140, 4140, C-Mo (Amola), and 1045. He concludes that the behavior of the steel under concentration of stress is a highly important factor and that, since the severity of the stress concentration varies with the design and the type of service, the steel, and its treatment, must be chosen in regard to those factors rather than selecting any one steel and laying down physical test requirements in the hope that they will be applicable to all designs.

Almen and Boegehold<sup>15</sup> made a very extended study of carburized rear axle gears and concluded that for this service the primary criterion for choice among alloy steels was not the tensile or impact properties, but rather the resistance to overload in fatigue. However, they could find no evidence that one alloy steel of those normally used was superior to another on that score. The choice, they concluded, should be governed by the warping tendencies, machining characteristics, and cost.

**Distortion, Cracking, Quenching Stresses.** A steel and a heat treatment that will avoid distortion will give quiet, long-lived gears. A coarse-grained steel is more likely to distort than a fine-grained one, so controlled grain size may be more important than alloy content.

Graves points out, in regard to non-carburized material, that quenching cracks are more prevalent in water-hardening steels. To avoid this and to minimize distortion, oil quenching is used; but this may not give sufficient surface hardness unless the hardenability is built up by alloying. We have emphasized the effect of alloys in regard to depth hardening, but shallow hardening is often required so that a warped piece may be straightened without cracking it, calling, therefore, for a soft core.

Boegehold<sup>2</sup> emphasizes the size limitations within which C steels will harden deeply enough to give strength, and the necessity for water quenching to give this sufficient depth of hardening. The water quenching is accompanied by danger of cracking in an irregularly shaped part. These limitations lead to the use of oil-hardening alloy steels for parts in which C steels would be unsatisfactory from these causes.

When hardness of the order of 400 Brinell is required, the low draw temperature that the C steel demands leaves residual quenching stress in the steel, which is likely to cause failure when added to the local stress concentration from the external loading at crucial points.

Distortion in hardening, particularly harmful in gears, is important in many other cases in the automobile where fit and alignment must be held within such close tolerances. The milder the quench, the less the distortion difficulty.

This same argument can, of course, be further extended, as Muir<sup>12</sup> has done, to favor the choice of highly alloyed air-hardening steels, with air quenching and tempering replacing oil quenching and tempering. Of course, the greater raw material cost of such steels and their lack of good machining qualities without special annealing limit their applicability for mass production parts.

Boegehold says that, beside ordinary tensile properties, which can be approximately matched in a variety of steels, the following count heavily in appraising steels for automotive service: critical quenching speed, i.e., depth-hardening ability, susceptibility to notches under repeated stress, distortion in hardening, and machinability at various hardnesses. Moore<sup>17</sup> brings out a similar point of view for aircraft engine service, as does Johnson.<sup>13</sup>

The whole discussion of competitive alloy steels emphasizes the fact that, while there is a tendency toward using the less highly alloyed, cheaper steels (especially those of controlled hardenability) in the automotive industry in general, first cost is often overbalanced by subtle distinctions that do *not* appear in the ordinary evaluation of the steels by tensile tests.

#### BIBLIOGRAPHY

1. LIPPERT, T. W.: Stainless Steel, *Iron Age*, v. 141, May 5, 1938, pp. 28-34.
2. BOEGEHOLD, A. L.: Plain Carbon Steel, *Metal Progress*, v. 31, February, 1937, pp. 147-152.
- BOEGEHOLD, A. L.: Advantages of Alloy Steels, *Ibid.*, v. 31, March, 1937, pp. 265-269.
- BOEGEHOLD, A. L.: Appraising a Steel for a Given Duty, *Ibid.*, v. 31, April, pp. 403-406.
3. PETERSON, O. N.: New Concepts of Metallurgical Control in Automotive Steels, *Steel*, v. 100, June 14, 1937, pp. 36-39, 72.
- PETERSON, O. N.: Metallurgical Control in Automobile Manufacture, *Heat Treat. & Form.*, v. 23, 1937, pp. 326-330.
4. GRAVES, W. H.: What Steel Is Best for This or That Part? *Metal Progress*, v. 29, April, 1936, pp. 36-42.
5. HALL, J. D.: Tool Steel Data Including Analysis: A Compilation of Outstanding American Tool Steel Manufacturers, 1937.

6. DAWE, C. N.: "The Steel (Physical Property) Atlas." Cleveland, Am. Soc. Metals, 1936, 87 pages.
7. GILLET, H. W., and E. L. MACK: "Molybdenum, Cerium, and Related Alloy Steels." New York, Chemical Catalog Co., 1925, 295 pages.
8. FRENCH, H. J.: Comparison of the Alloying Elements Chromium, Nickel, Molybdenum, and Vanadium in Structural Steels, *Trans. Am. Soc. Steel Treat.*, v. 11, 1927, pp. 845-884, 1000.
9. CLAPP, W. H.: The Relation of Mechanical Testing to Steel Performance. *Am. Petroleum Inst.*, 17th Annual Meeting, Chicago, November 10, 1936. 8 pages.
10. GROSSMANN, M. A.: "Principles of Heat Treatment." Cleveland, Am. Soc. Metals, 1935, 147 pages.
11. THUM, E. E.: Izod Toughness vs. Tensile Strength of Heat Treated S.A.E. Steel, *Metal Progress*, v. 29, May, 1936, p. 57.
12. STRAUSS, J.: Medium Carbon Pearlitic Manganese Steels, *Trans. Am. Soc. Steel Treat.*, v. 14, 1928, pp. 1-26.
13. JOHNSON, J. B.: Aircraft Engine Materials, *S.A.E. Journal*, v. 40, April, 1937, pp. 153-162.
14. KNOWLTON, H. B.: Physical Properties of Axle Shafts, *Trans. Am. Soc. Metals*, v. 25, 1937, pp. 260-296.
15. ALMEN, J. O., and A. L. BOEGEHOLD: Rear Axle Gears: Factors Which Influence Their Life, *Proc. Am. Soc. Test. Mat.*, v. 35, part 2, 1935, pp. 99-146.
16. MUIR, J.: Tension and Notched-Bar Tests on an Alloy Steel, *J. Royal Tech. College*, Glasgow, v. 4, January, 1937, pp. 1-11.
17. MOORE, R. R.: Materials of Construction in Aircraft Engines, *Proc. Am. Soc. Test. Mat.*, v. 30 part 2, 1930, pp. 88-98.



### SECTION III. THE SPECIAL-PURPOSE STEELS

#### CHAPTER 19

#### STEELS FOR HIGH- AND LOW-TEMPERATURE SERVICE

**Requirements of High-temperature Service.** The power-plant industry utilizes high pressure superheated steam in the range 700–1200° and requires steel for boiler tubes, valves, piping, superheaters, turbine rotors, blades and casings, and in bolting stock. The bolts do not reach full steam temperature because they are on exterior flanges.

In the petroleum industry, cracking still tubes operate with an oil temperature of 800 to 1200° with outside tube wall temperatures of 1100–1500°, while hangers and furnace parts run as high as 1700–1800°.

Heat-treating furnaces require conveyors, carburizing boxes, radiant tubes, etc., operating at 1500–1900° or over. Chemical industries require vessels for high temperatures and pressures, and roasting furnaces for the metallurgical industry have peculiarly severe requirements.

In many of these services some type of corrosion must be resisted, while the parts that come into contact with air or flue gases must resist oxidation and, in the latter case, the actions of sulphurous gases as well. The service conditions to which steels are subjected at elevated temperatures may be extremely varied. Evaluation of a steel must therefore be made in the light of the service for which it is destined.

Some, or all, of the following questions must be answered in the selection of a steel for high-temperature service:

Will it withstand oxidation or corrosion under the environment of service?

Will its metallographic structure, and hence its properties, change? That is, what is its stability? Will it become brittle?

How will it behave under momentary overload at the normal operating temperature, or under normal load at increased temperature for a short period?

What is its load-carrying ability under the normal temperature of service?

**Oxidation and Corrosion.** The oxidation and corrosion problems can be answered only by specific tests or experience under service conditions. Under oxidizing conditions, in air or flue gas, for example, C steels scale so heavily at temperatures above  $1100^{\circ}$  that, when really long life is necessary, some alloy must be added to increase the scale resistance. Such an addition may be advisable for service near to or above  $1100^{\circ}$  on the score of scaling alone, and most authorities set  $900$  to  $1000^{\circ}$  as the operating limit for C steels. Sulphur-containing gases introduce an especially serious corrosion problem. The elements Cr and, to a smaller extent, Si and Al, are introduced for resistance to oxidation and scaling. The same elements are advantageous in resisting various types of corrosion.

**Types of Heat Treatment That Are Applicable.** Continued high-temperature service naturally softens a quench-hardened steel, having the effect of a very prolonged draw, so that strength conferred by quenching can normally be utilized only in such relatively low-temperature services as that of bolting stock. In that service the corrosion-resistance requirements are seldom severe, so quenched and tempered alloy steels of the constructional steel class may be utilized.

An attempt to stabilize the structure is almost invariably made by giving the steel a final treatment, before service, at a temperature above that of service. Complete stabilization would involve not only annealing but also spheroidization, either of which sacrifices strength. Hence, for service at those temperatures where pearlitic steels may be utilized, the steels are commonly made as stable as possible without annealing, by normalizing and then drawing at a temperature below  $A_{c1}$ , but well above the operating temperature. Although the spheroidized, annealed state is the most stable, the load-carrying ability of such a microstructure is inferior to that of lamellar pearlite, and, in practice, the lamellar structure, in a suitably alloyed steel, may be sufficiently stable to resist spheroidization and consequent loss of load-carrying ability.

**Precipitation Hardening and Stability.** As has been mentioned in Chap. 8, Vol. I, and will be discussed further in the next chapter, carbide separation, by a precipitation-hardening type of phenomenon, has always to be reckoned with, especially in the stainless and heat-resisting alloys and in those of lower Cr content, if, in service, the alloy is heated and cooled through the temperature ranges in which changes in solubility of carbides occur. This holds unless the C has been bound as an insoluble carbide, or unless, when the carbides are precipitated from solution, they can be induced to separate throughout the grains instead of in the grain boundaries. If they come out in the grain

boundaries, the alloy may become more brittle, though not necessarily so at the precipitation temperature. Brittleness may be noted only at room temperature.

The temperature at which the carbide goes into solution, and the lower temperature range in which it precipitates from solution, will vary with the nature of the carbide and of the matrix. Operating temperatures are often in the precipitation range.

**The Carbide Balance.** Carbides are naturally brittle, and a carbide envelope about the grain boundaries lends brittle behavior to a steel whose grains themselves may be highly ductile. Carbides of small size, uniformly distributed, or the carbide in a pearlitic structure, give strength. In C steels the high-temperature strength is higher as the C increases, up to perhaps 0.40%, but does not rise very rapidly thereafter. At very high temperatures the C content is much less of a factor in strength than it is at normal temperatures. Thus a carbide balance is required. Low-C, spheroidized carbides, or inert carbides (of Ti and the like), which act like spheroidized carbides, do not supply strength, while considerable amounts of C that can go in and out of solution may be fatal to ductility.

**Embrittlement.** Dixon<sup>1</sup> has pointed out that the low-C, 5% Cr steels, without Mo or W, may become embrittled in oil-still tube service, fracturing at room temperature when subjected to impact in the process of removing coke that has deposited in the tubes, although the embrittled metal is tough at operating temperature. This is analogous to any ordinary temper-brittleness phenomenon and does not seriously affect the properties other than impact, and that only when the steel is cool. Dixon prefers to use 0.50% Mo, the usual remedy for temper brittleness, to avoid this, though there are some who advocate W. Babcock and Wilcox add 1.5% Mo to a 9% Cr tube steel that they may be sure to have enough to avoid embrittlement.

In the highly alloyed austenitic heat-resisting alloys, elevated temperature service within the precipitation zone of temperatures, or even a short sojourn at such a temperature, may drop the elongation to a low figure. If the concentration of carbides in the grain boundaries finally leads to selective oxidation there, the alloy may become full of tiny intercrystalline cracks. Tapsell<sup>2</sup> has discussed this in some detail. He has even reported it in C steels under load at the very high temperatures, for such steels, of 1100–1200°. The actual cracking seems to be associated with high deformation, and some designers consider that loading in high-temperature service should be kept down so that, within the life of the part, no more than an extension of the order of 1% can be reached, irrespective of how much deformation the part would stand without ceasing to function. Yet this is not necessarily a

valid limitation. Cross<sup>3</sup> loaded an 0.07 C, 18 Cr, 8% Ni steel at 1200° for 11,500 hr. at 8345 lb./in.<sup>2</sup> which produced the large elongation of 3.3%. On another specimen the same load and temperature for 18,000 hr. produced 4.5% elong. The original material, not loaded or heated, showed 104 Izod. The deformed sample gave, after 11,500 hr., 54, after 18,000 hr., 34. There was no sign of intercrystalline cracking. At this rate, the service life would be long before this particular lot of material would have become seriously embrittled.

Bailey<sup>4</sup> considers that all low-alloy steels of high load-carrying ability in long-time, high-temperature service will fail, under excessive load, by intercrystalline cracking, with comparatively low ductility. He advocates that the pearlitic steels containing Mo should not be stressed high enough to exceed 0.5% extension in their period of service. Aborn and Miller<sup>5</sup> point out that such failures are met only at extreme loads, producing a rate of some 20% or more extension in 10,000 hr. Such extreme loads were used in the laboratory tests from which Bailey drew his conclusion.\* No one would ever apply such high loads in service. Aborn and Miller tested a 0.50% Mo, low-C steel under 4,000 lb. in.<sup>2</sup> stress at 1100° for 3,000 hr., again with no sign of intercrystalline cracking. Bailey<sup>6</sup> now says that intercrystalline cracking is normal behavior, not at all to be feared if properly allowed for in design. White, Clark, and Wilson<sup>7</sup> stressed a variety of steels at very high loads in order to see how long they would hang on without rupture at elevated temperatures. They met intercrystalline cracking in low-C steel at 12,000 lb./in.<sup>2</sup> at 1000°, failure occurring in 1550 hr. with 10% extension. This load is several times the "allowable working stress" for that steel and temperature. Aborn and Miller conclude that loading in a test, far above service conditions, does not permit the steel to adjust itself by the relatively slow structural rearrangement which, under service conditions, permits flow without cracking. In discussion of the paper by White, Clark, and Wilson, Whitney stated that C-steel tubes, in actual operation in the oil industry, under allowable design loads at 1000°, last 7 or 8 years; and, when they do become thinned by corrosion so that they fail, they deform before failure rather than fail with a brittle, intercrystalline fracture. Yet some 18 : 8 tubes are reported by Dixon<sup>1</sup> to have failed in service, in brittle fashion, though many other 18 : 8 tubes have given exceptional service.

Taprell<sup>2</sup> remarks that some cases of intercrystalline failure of heat-resisting alloys have been at temperatures that might be considered above the economic temperature range for the materials.

Although most cases of intercrystalline brittleness thus relate to ex-

\*Thielmann and Parker<sup>93</sup> obtained intercrystalline failure of CMo in 964 hr. at 1100°, under the load, very high in relation to design loads, of 10,000 lb./in.<sup>2</sup>

cessive temperature or excessive load, above that contemplated in reasonable design, the propensity toward such brittleness is a factor to be considered. Testing at excessive temperature or load may lead to the conclusion that danger is to be feared, when the conclusion may be quite unwarranted in actual service.

**Stability vs. Load-carrying Ability.** Even though no brittleness may be brought about by a structural change produced through long sojourn at high temperature, changes may occur that lower the load-carrying ability. As has been mentioned above, lamellar pearlite has better load-carrying ability than a structure in which the carbide is spheroidized. Sorbite spheroidizes more readily than pearlite, so that quenched and tempered sorbitic steels are generally applied only in relatively low-temperature service. At the temperatures where long exposure would soften sorbitic steel, either annealed steels, or those normalized and drawn well above the temperature of service, are generally chosen. The stability of the normalized and drawn material needs to be established for each steel and set of conditions.

**Overload.** Ability to stand momentary overload or brief periods of increased temperature, is needed when service conditions are not under good control. This ability may be measured by the short-time tensile test at high temperatures. Methods for this have been described by the A.S.T.M.<sup>5</sup> Figure 1 shows results obtained by a subcommittee of the Joint Research Committee and clearly brings out how rapidly increase in temperature drops the strength of C steel in the range 600–1100°. Although the short-time test is essential for evaluating resistance to overload, it unfortunately does not indicate the load-carrying ability for an extended period.

**Proportional Limit.** Under long-continued loading, permanent plastic flow occurs at high temperatures, so slowly that the stretching is called "creep." As the loads are decreased, to the order of those that can be borne in service, it requires measuring equipment of great sensitivity to detect and measure the creep, as well as a long period in which to allow readily measurable deformation to accumulate. Measurements of "proportional limit" in the short times of ordinary tensile tests carried out at high temperatures bear no sure relationship to creep.

Kanter and Spring<sup>6</sup> long ago showed (Fig. 2) that proportional limit determinations bear no direct relationship to creep values. The matter was examined by an A. S. T. M.-A. S. M. E. Subcommittee,<sup>10</sup> which reported that the test had too many uncertainties in itself and that evidence of identity of action between deformation in such a test and actual creep was lacking. Clark and White<sup>11</sup> show, as did Kanter

and Spring, that proportional limit and creep curves cross. At the lower temperatures, when strain hardening is marked, proportional limit gives a lower stress; and in the higher temperature range, where plastic flow is marked, the proportional limit gives a higher stress. The temperature at which the curves cross, and the angles at which they cross, vary with different steels. Hence no sure prediction of load-carrying ability can be deduced from proportional limit figures.

**"Time Yield."** Hatfield <sup>12</sup> has suggested a "time-yield" test which might be considered as a moderately slowly made proportional-limit

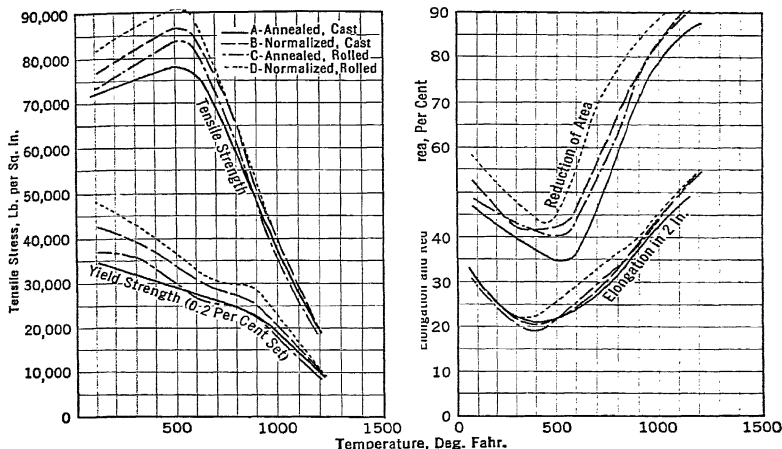


Fig. 1.—Short-time, high-temperature tensile tests of cast and rolled C steel. 0.28 C, 0.33 Si, 0.66% Mn, annealed at 1600°, or normalized from 1650°.

test. White and Clark <sup>13</sup> studied this in comparison with creep curves and found that the curves cross, much as the proportional limit and creep curves do, and that the time-yield values cannot be directly converted into creep values.

Any of these short-time tests give information that is of some interest, and if there is complete certainty that the whole high-temperature behavior of a steel will be the same as that of another if the two agree in *one* respect, such tests might have value as acceptance tests. However, the assurance of complete agreement is never certain till *all* the tests have been made. The chances that such tests will mislead are all too great.

**Time to Rupture Tests.** A method has been suggested by White, Clark, and Wilson<sup>7</sup> in which a high load is applied at a given high temperature and the time for rupture noted. Then a lower load is applied to another specimen, and so on. The times of rupture and the loads are then plotted on log-log scales and extrapolated to 100,000 hr. The load for rupture in 100,000 hr. should of course be considerably higher than the load, determined by extrapolation of creep curves, for an extension of 1% in 100,000 hr., unless one has reason to suspect that there is almost entire loss of ductility so that the steel could not stretch 1% without breaking. These authors give the following data for tests at 1000° on steels annealed at 1550°.

Mn	Cr	Mo	Grain Size	Extrapolated Stress for Rupture in 100,000 Hr. lb. in. <sup>2</sup>	Extrapolated Stress for Creep of 1% in 100,000 Hr. lb. in. <sup>2</sup>	Elong. of Short-time Tensile Test	Elong. in Rupture Test
0.15	0.23	0.50	4-5	2,800	2,700	42.5	12 at 10,000
0.07	0.72	0.42	4-5	9,000	13,000	32.5	27 at 5,000
0.10	1.55	0.38	4-6	10,000	5,500	43	56 at 1,000

No relationship is shown between the extrapolated values. The extrapolations indicate that the second steel, although it is not losing ductility rapidly, would break at a load far below its indicated creep strength, which appears to be an absurd conclusion. As the discussion of this paper indicates, the "rupture strength," while an interesting value, is not proved to have any direct relationship to load-carrying ability at reasonable design values and certainly cannot be taken as a substitute for creep tests, especially since, as discussed above, intergranular embrittlement that is not representative of the material under service conditions may be met in such overload tests. A modification of such a method is to determine the load for a given time to complete fracture. Pilling and Worthington<sup>14</sup> show in Fig. 3 that neither such a method nor the short-time tensile test has direct relationship to long-time, load-carrying ability in Fe-Ni-Cr alloys.

Tests to destruction seem too far removed from the property actually sought, which is the ability to withstand load, not only without fracture but also without more than a very limited deformation. Direct determination of the desired property is the only sure and straightforward method of evaluation. The time-to-rupture tests and others of that class, are generally advocated by those who have developed the tests, merely as tests for identity with a steel, taken as a standard, whose creep properties are known. Whether such identity tests are sufficient, remains to be seen.

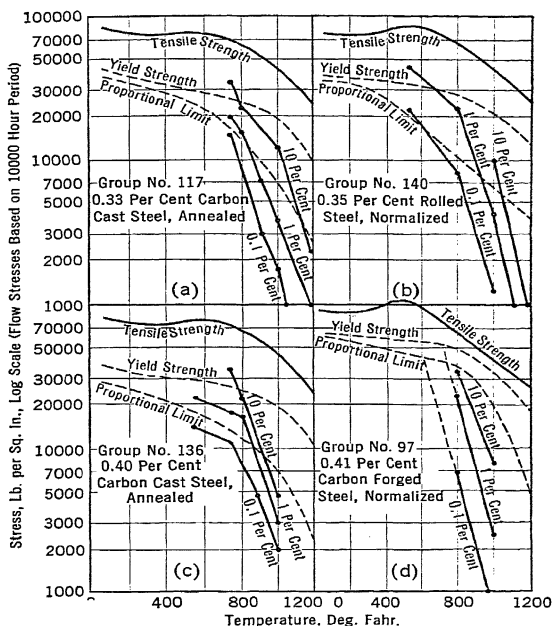


Fig. 2.—Stress-temperature curves (semi-log plot) for short and long time high-temperature tests. (Kanter and Spring.<sup>9</sup>)

Note that the creep curves cross the proportional limit curve and bear no fixed relation to it.

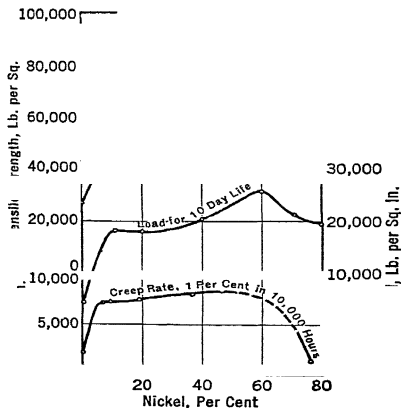


FIG. 3.—Effect of Ni content on strength at 1200° of wrought alloys with 20% Cr. The balance is Fe. (Pilling and Worthington.<sup>14</sup>)



**Creep.** Thus we come to the necessity of considering actual creep tests, i.e., long-time loading tests in which plenty of time is allowed for plastic action to manifest itself. The loads are not greatly above design values, and very small deformations are measured. The value sought is the load, which, at the given temperature, will produce a fixed deformation in a fixed time. The values generally used are those loads which will produce a rate of elongation of either 1% in 10,000 hr., or 1% in 100,000 hr. These two different criteria are used because the permissible percentage deformation varies with the type of service and the expected service life. In furnace parts that do not demand close clearance, as high as 5% total deformation in 10,000 hr. might be

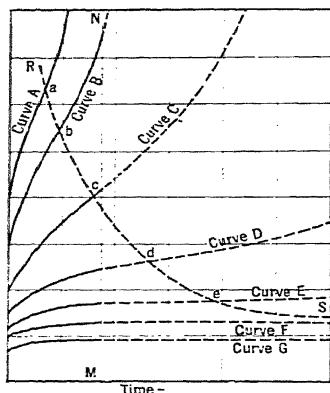


FIG. 4.—Schematic plot of typical creep curves for different stresses, ranging from very high load, Curve A, to very low load, Curve G, at a constant temperature. (McVetty.<sup>15</sup>)

borne, provided that such deformation can be had without embrittlement. Most oil cracking and similar uses, excluding turbines, might stand 1% in 10,000 hr. (This is better but less commonly expressed as 0.1% in 1,000 hr., or equivalent in a longer test, e.g., 0.2% in 2,000 hr., to indicate that the test ran only that long, and to avoid extrapolation in the reported figure.) Turbines, however, are expected to have a useful life of 20 to 25 years, so 1% in 100,000 hr. or 0.01% in 1,000 hr. represents excessive distortion because of the close clearance of turbines. Bailey<sup>6</sup> is especially interested in deformations of 0.1 to 0.3% in 100,000 hr. in studying steels for turbines. The tests are usually run for not less than 1,000 hr., but seldom more than 3,000 hr. (except in certain fundamental studies), and the results are then extrapolated.

A specimen that creeps at the rate which gives a total extension of 1% in 10,000 hr. is extending only at the average rate of a millionth

of an inch per inch per hour. Great precision in measurement and in control of temperature is therefore required. The effect of temperature on creep rate is very much greater than it is on short-time, high-temperature tensile strength, as Fig. 2 shows. At a given temperature, increase in load also increases the deformation very markedly. This is shown diagrammatically in Fig. 4, after McVetty.<sup>15</sup> In Fig. 4 line *NM* represents the usual test period of say 1,000 hr. The real rub comes in extrapolating beyond the test period. Curve *RS* of Fig. 4 is drawn

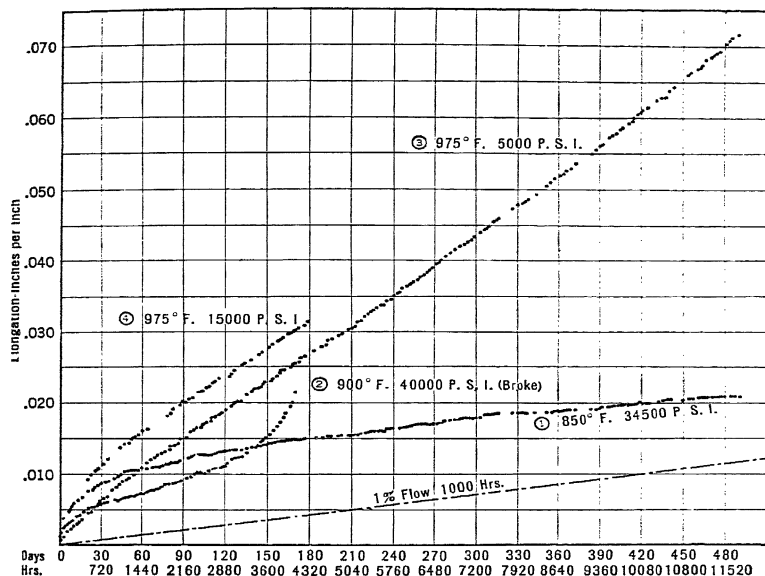


Fig. 5.—Creep curves for four specimens. Two of these were under test for 16 months. (Foley.<sup>14</sup>)

through the inflections of the family of curves. If we could construct *RS* with certainty from the test data, extrapolation would be easy; but in most materials under usual test conditions, the upturn, shown in Curves *A* to *D* of Fig. 4 occurs, as in Curve *E*, far beyond the usual test period. With such a family of curves we may be quite sure that any upturn in Curve *F* would come only after a very long period.

Foley<sup>14</sup> shows data from long-continued creep tests in Fig. 5. These are on four different analyses of low-C, low-alloy, pearlitic steels. Steel

No. 1, at 700 hr., was stretching more than steel No. 2; but at about 1000 hr. the curve for No. 1 started to flatten out, while at 2000 hr. that for No. 2 started to turn up. Steel No. 3 maintained a constant rate from the start. It will be noted that the loads and temperatures in this comparison varied widely. However, it is plain from Fig. 4 that the initial slope of the curve may be quite misleading as an index of its future course, so that one needs a whole family of curves like that of Fig. 4 in order to see what behavior may reasonably be expected at different loads or different temperatures.

That steels in a stable condition may maintain a uniform rate of creep for very extended periods is shown by Figs. 5, 6, and 7, which represent some of the longest tests on record.

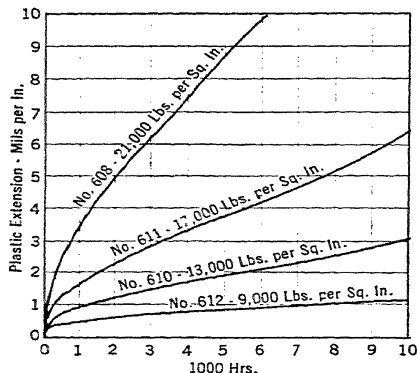


FIG. 6.—Creep curves for 10,000-hr. test of Ni-Cr-Mo steel annealed at 1515°, tested at 857°. (Clark and Robinson.<sup>53</sup>)

0.37 C, 0.16 Si, 0.55 Mn, 1.78 Ni, 0.77 Cr,  
0.36% Mo.

**Strain Hardening vs. Annealing.** When the creep curve shows a consistently decreasing rate, as does Curve 1 in Fig. 4, it is considered that strain hardening is occurring at the load and temperature used and that this is strengthening the steel, so that the extremely tiny reduction in cross section, occurring as the steel stretches, is overbalanced. When the curve turns up, as does Curve 2 in Fig. 5, it is considered that any strain hardening is being overbalanced, i.e., the conditions are those of annealing, so that there is insufficient opposition to plastic flow. Obviously, there is an in-between condition in which the annealing tendency is just sufficient to balance the tendency toward strain hardening, and the creep curve remains straight for long periods. (See Fig. 7.) If there is assurance that it will remain straight and that the total deformation at that creep rate will not exceed the permissible deformation, in the life of the part, design could use that load at that temperature. In turbine design there is no fear that failure will ensue.

# PLOT OF 20,000-HR. CREEP TEST

347

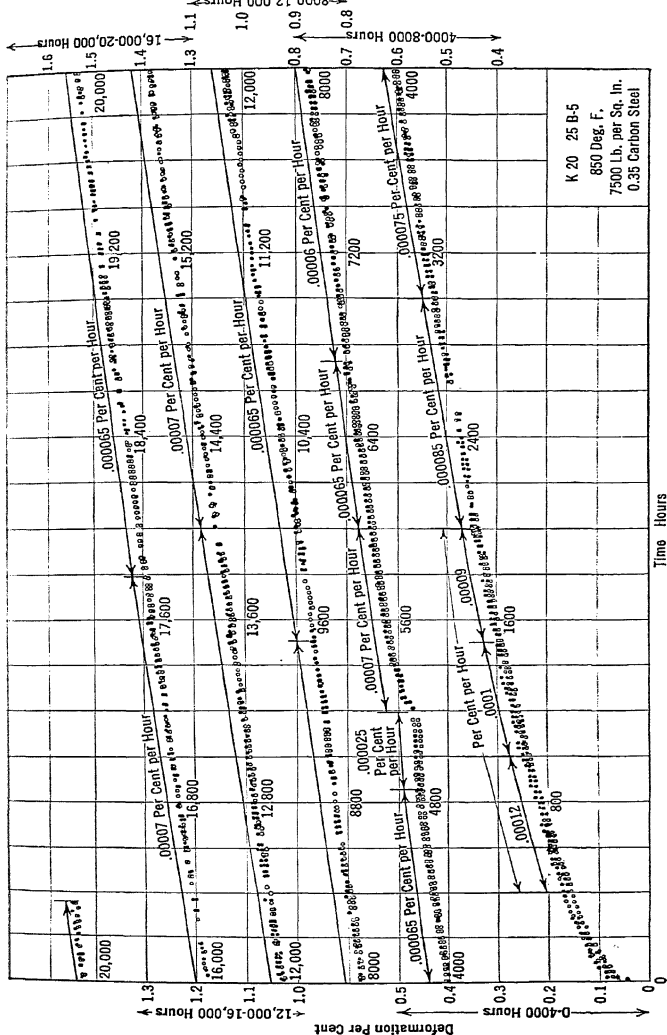


Fig. 7.—20,000 hr. creep test on annealed C steel of 0.35 C, 0.19 Si, 0.55 Mn, 0.03 S, 0.016% P. (Cross and Lowther.<sup>81</sup>) Austenitic grain size, 7.8, A.S.T.M. Creep rate at 4000 hr. equivalent to 1% in 10,000 hr., Actual extension in 10,000 hr. 0.9%; in 20,000 hr. 1.53%. The creep rate diminished for 4000 hr., and remained constant for the next 14,000 hr. See Fig. 12 for further test on this specimen.

The limiting feature of design is the deformation which might make the turbine rub against its casing in the some 20 years of its expected service.

In general, one feels safer if the load is such that a curve is produced of the strain-hardening type, like No. 1 in Fig. 5. It may be

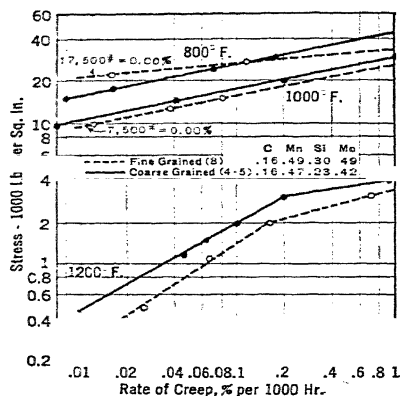


FIG. 8.—C-Mo steel, annealed at 1550°.

Grain size determined by McQuaid-Elliott test at 1700°, not at the temperature actually used.

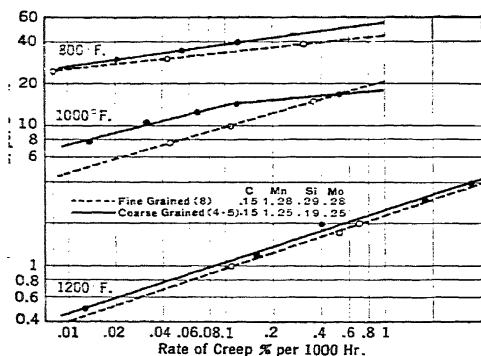


FIG. 9.—Mn-Mo steel, normalized from 1725°, drawn at 1200°.

Note that the slopes of the curves are not necessarily the same, and that the log-log plot is not necessarily a straight line.

FIGS. 8 AND 9.—Influence of grain size on creep curves at 800, 1000, and 1200°. (White and Clark.<sup>17</sup>)

impossible to find an alloy that will strain harden at the desired operating temperature. Heat-resisting alloys of the type chosen for use at 1800–2000° seldom, if ever, show strain hardening, at such temperatures, but their uses seldom require close adherence to dimensions. The

relation of creep to design in such cases is chiefly to ensure that the loading is such that the deformation, in the desired life of the part, does not become so large that brittle failure might ensue. In this the stability of the alloy and its behavior as to carbide precipitation are as important as actual creep rate.

**Creep Rate vs. Stress.** The effect of increase in stress upon creep rate at a given temperature is very large indeed. At a given temperature an increase in stress increases the rate in exponential fashion. That is, the equation of the creep curve can be written  $y = mx^n$ , where  $y$  is creep rate,  $x$  is stress, and  $m$  and  $n$  are constants. This means that if creep rates are logarithmically plotted against stress, and the equation

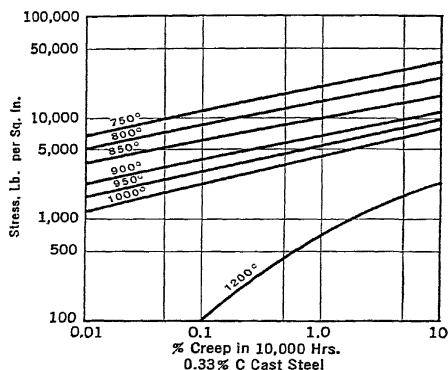


FIG. 10.—Log-log plot of relation of stress to creep at various temperatures. (Kanter.<sup>18</sup>)

is obeyed, the curve is a straight line. In the majority of cases the curves so plotted are straight, which means that interpolation between observed points can be safely made and that some degree of extrapolation may be justified. However, the log curves are not necessarily straight, as Figs. 8 and 9 from White and Clark<sup>17</sup> and Fig. 10 from Kanter<sup>18</sup> show.

If the slope of all logarithmically plotted creep curves were the same so that one could determine one point and draw a line through it at the standard slope, both testing and extrapolation would be facilitated; but Fig. 8 shows that the slopes may vary. Figure 11 also shows this. As Norton<sup>19</sup> states, expressed in terms of the exponential formula,  $n$  averages around 5, but may vary from 3 to 17.

**Creep Rate vs. Temperature.** The effect of change in temperature at a given load is also very great. Figure 12 shows a 5 to 1 increase in rate for the particular steel and load used as the temperature is

## 350 STEELS FOR HIGH- AND LOW-TEMPERATURE SERVICE

increased from 550 to 900° and a 5 to 1 decrease when it is decreased from 850 to 800°. Norton comments that the rate of creep is, in some cases, doubled for a rise of 25°. This rate of change cannot be exactly predicted for all steels and loads. Bailey and co-workers<sup>20</sup> give the approximate relation for C steels in Fig. 13.

Robinson<sup>21</sup> discusses the effect of temperature on creep and points out that different steels show different degrees of sensitivity to increase in temperature. He makes a statement that brings out clearly the

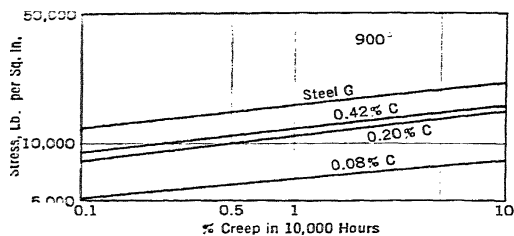
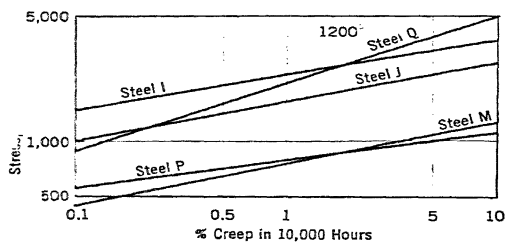


FIG. 11.—Log-log curves for creep of various steels. (After Norton.<sup>19</sup>)



Note the differences in slopes.

Steel	Composition $\%$							Steel	Composition $\%$						
	C	Si	Mn	Cr	Ni	W			C	Si	Mn	Cr	Ni	W	
G	0.10	0.86	0.31	17.60	0.23	...		J	0.20	0.36	0.80	26.94	0.51	...	
Q	0.25	0.25	0.57	0.38	0.22	1.95		M	0.40	0.20	0.59	0.66	3.39	...	
P	0.37	3.51	0.35	2.25	0.22	...		P	0.40	0.25	2.22	...	...	...	

practical result of increase in temperature in decreasing creep resistance. By assuming that a steel of a given sensitivity is used in the construction of a turbine designed to operate at 750°, and to show only such change of dimensions, due to creep, as will be permissible during its normal life, it may be calculated that, if the temperature is allowed to rise to 777°, each hour of running at that temperature "ages" the turbine, in respect to dimensional changes, as much as 3 hr. of normal operation. That is, the turbine that would be good for 20 yr. at 750°, would be scrap, or due for a major overhaul, in 7 yr. at 777°.

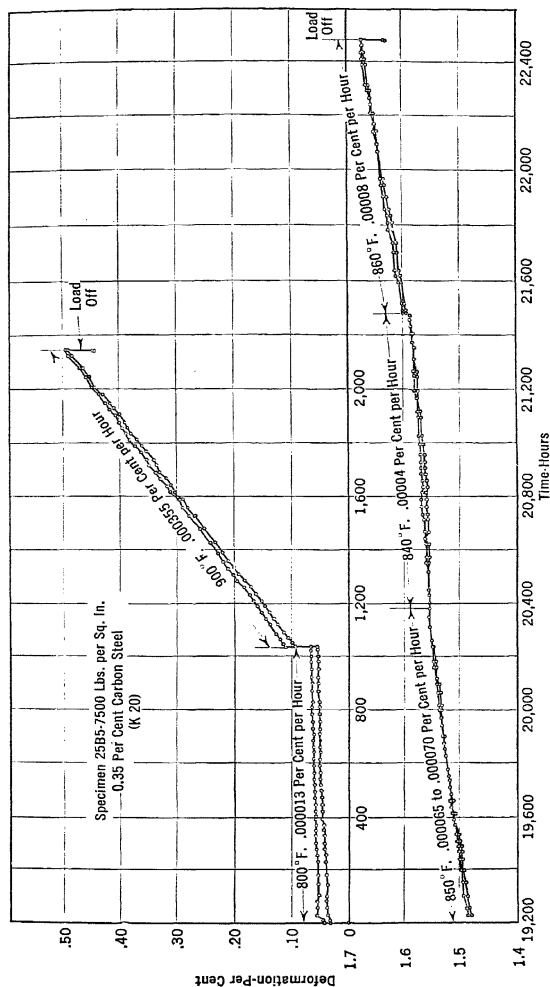


Fig. 12.—Rates of creep on annealed, fine-grained 0.35% C steel at different temperatures. (Cross and Lowther.<sup>72</sup>)  
The curve for 850° is the last 1000 hr. of the test shown in Fig. 7. This steel showed no embrittlement after almost 25,000 hr. under load at the temperatures shown, and after a total elongation of about 2%.



**Creep-testing Procedure.** The great variation in creep rate with temperature requires extreme precautions as to constancy of the test temperature and uniformity of temperature over the gage length, in creep testing. This has been discussed in detail.<sup>22</sup> The extensions to be measured are so tiny that great precision is necessary. Long gage lengths facilitate this (though introducing increased difficulty in maintaining temperature uniformity) as does a long period of creep, in which to accumulate easily measurable deformation. Even more important to length of test is the fact that it is necessary to allow the strain hardening or the annealing tendency, whichever predominates, to

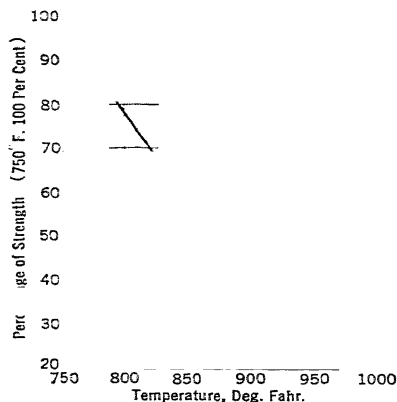


FIG. 13.—Relative creep resistance of medium carbon steels with increase in temperature. Resistance at 750° is taken as 100%. (Bailey and co-workers.<sup>20</sup>)

manifest itself clearly. Whereas some foreign investigators draw conclusions from readings obtained in the first hundred or first few hundred hours, the creep data subcommittee<sup>23</sup> of the Joint Research Committee found it necessary to discard all creep data save those extending to at least 500 hr., and preferably to at least 1,000 hr. Figure 5 is ample justification for declining to draw conclusions from abbreviated creep tests.

These requirements of precision in temperature control and of long time of test, together with the expensive equipment necessary (Figs. 14 and 15), make the accumulation of reliable creep data slow and costly. Many statements in the literature, especially the foreign literature, are made as to the effect of some alloying element or combination of elements, based on short or "accelerated" tests, or on early tests made before the necessity for close temperature control was realized. Such statements have to be greatly discounted. Credence, however, can be given to practically all the creep data now being obtained in the United



FIG. 14.—Battery of creep test units. (Battelle.)

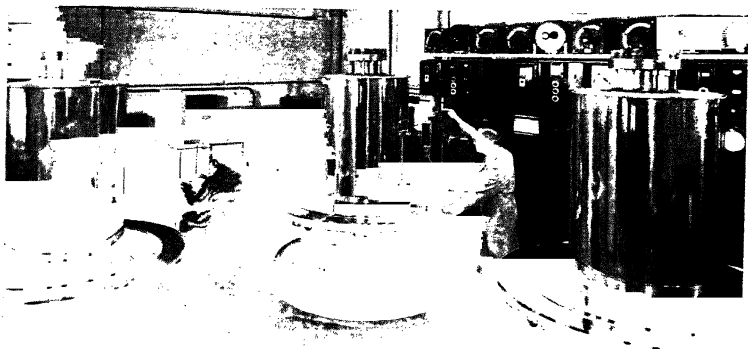


FIG. 15.—Three multiple creep test units. (Westinghouse El. & Mfg. Co.). Each unit can test a dozen specimens at once. The outfit is described by McVetty.<sup>88</sup>

States and to much foreign data, like those from the National Physical Laboratory of England.

The "time yield" and other short-cut methods that attempt to draw conclusions as to ultimate rate of creep from the early stage of the

creep curve are handicapped by the fact that the curve is changing slope in that period. Wilson<sup>24</sup> gives a good explanation of the phenomena of this period. He attributes the shape of the curve in the initial stage of creep to the fact that the individual grains of steel would show different properties if they could be tested in different directions, so the behavior of a grain depends on the inclination of its axis to the direction of application of stress. "Some grains in the random grouping across any cross-section will be more favorably situated than others, and the stiffer ones will carry the load at first, but as they deform, the load will be distributed among some others. The net effect

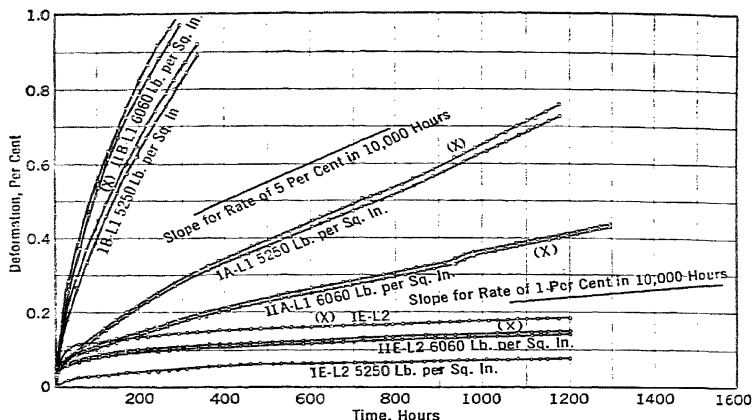


FIG. 16.—Time-deformation curves at 1100° for 5 Cr 0.50% Mo steel specimens cut from still tubes. (Cross and Johnson.<sup>25</sup>)

Longitudinal specimens. Steel I, 0.14% C; Steel II, 0.18% C. Treatment A, slow cooled from 1550°; reheated 6 hr. at 1425°. Treatment B, normalized from 1550°; reheated 6 hr. at 1425°. Treatment E, slow cooled from 2100°. X signifies the inner surface of the still tube.

is the same as though the material were actually becoming stronger, with the result that the amount of creep decreases. At the same time, the plastic deformation increases strength to some degree and thus contributes toward decreasing creep."

Obviously, different steels will behave differently as to the rate at which this stiffening effect occurs; they will take varying times in which to "make up their minds" as to the rate they will creep under the test conditions. (Compare the upper curves in Fig. 17.) Hence it is imperative that ample time be given for a representative creep rate, and not merely the more or less accidental initial creep rate, to manifest itself. Unless the steel is so weak at the test load and temperature that

it extends so fast that even in a short time it is obvious that the steel is not suitable to resist those conditions, modern creep testing considers that the first 500 hr. is merely an initial period of adjustment, and draws its main conclusions from the behavior after 500 hr. Short tests may, of course, serve as a rough means of sorting out obviously unsuitable steels from those that deserve extended tests as in Figs. 18-21. Precise temperature control, accurate load measurement, and suffi-

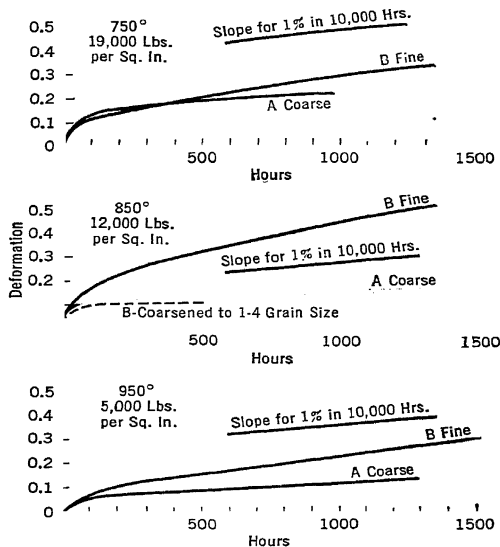


Fig. 17.—Creep behavior of coarse- and fine-grained specimens cut from heavy valve forgings. (Cross and Dahle,<sup>26</sup> Cross and Lowther<sup>27</sup>)

Steel	Composition %							Austenitic grain size, as normalized from 1550°
	C	Mn	Si	Ni	Cr	Mo		
A	0.27	0.63	0.18	0.04	....	tr		5-6
B	0.37	0.63	0.24	0.18	0.11	0.02		7-8 1-4, normalized after 2 hr. at 1910°

ciently extended periods of test are now so generally adopted that serious variations in results due to faulty laboratory technique are seldom met in results of creep tests made with modern precautions.

**Creep vs. Grain Size.** A matter that is still serious, in obscuring the real effect of alloying elements, is the effect of grain size upon creep. That coarse-grained steels are superior in creep resistance was stated in 1930 by Kanter and Spring,<sup>9</sup> especially in reference to cast *vs.* wrought steel, since cast steel ordinarily has a larger grain size than a similar

wrought steel, owing to the higher heating temperature of annealing or normalizing the cast steel. Cross and Johnson<sup>25</sup> (Fig. 16) showed large differences in creep of 5 Cr, 0.50% Mo steel with different heat treatment. The higher temperature treatment (*E*, lower curves Fig. 11) that resulted in large grain size gave high creep resistance and, naturally, low impact. Cross and Dahle,<sup>26, 27</sup> compared two forged steels, in specimens cut from large valves that had been normalized from 1550°. Steel *A* (Fig. 17) was coarse grained, steel *B* fine grained. No other feature would account for the differences in creep rates shown by these steels. When *B* was coarsened<sup>27</sup> by heating 2 hr. at 1900° before normalizing it showed the same creep rate as *A*. Note that the superiority of the coarse grain holds at 750° as well as at higher temperatures.

Clark and White<sup>28</sup> (Figs. 8 and 9) also showed that coarse-grained steels are, in general, superior in creep at high temperature, but found this did not hold on their 0.50% Mo steels tested at 800°. They suggested that coarse grain is superior in creep above the "equicohesive temperature" and poorer below it. The equicohesive temperature is a vague phrase used to denote a range, below which fracture is supposed to take place through the grains, and above which it is supposed to take place between them. By this criterion, this temperature is difficult to locate exactly. It is also supposed to mark the limit below which strain hardening can predominate and above which annealing occurs, i.e., it is the lowest temperature of recrystallization. Clark and White<sup>28</sup> indicate that, for C steel, the equicohesive temperature is about 850°. Jares<sup>29</sup> puts it higher, at 990°. Wilson<sup>34</sup> has recently placed it at 650–750° for low-C steels.

White, Clark, and Wilson<sup>30</sup> once remarked that, in the important range of industrial temperatures, the coarse-grained steels are not enough stronger to justify a preference for them. On the basis of later data it appears that this conclusion may need modification.

**Killed vs. Rimmed Steels.** It has long been known, as Kinzel<sup>31</sup> points out, that rimmed steel is inferior in creep resistance. Jenkins, Tapsell, Mellor, and Johnson<sup>32</sup> compared a rimmed steel at 840° and about 18,000 lb. in.<sup>2</sup> load with similar killed steels. The rimmed steel crept 400 times as much, in about 500 hr., as did good killed steels of the same C content. White and Clark<sup>28</sup> made a similar comparison, in terms of load for final creep of 1% in 10,000 hr., the killed steel giving 16,750 lb. in.<sup>2</sup>, the rimmed but 10,000. Rimmed steel seems to be definitely ruled out for high-temperature service.

**Grain Size in Killed Steels.** If we confine ourselves hereafter to killed steels, we find data for annealed C steels on record, indicating

that at  $850^{\circ}$  the load for a final creep rate of 0.1% in 1000 hr. may vary from the 7,500 lb./in.<sup>2</sup> of the steel in Fig. 7 to as high as 23,500 for a steel of larger grain size. However, by coarsening the grain of the steel of Fig. 7, by heating 8 hr. at  $1900^{\circ}$  and normalizing, the creep rate of this steel has been improved so that it approximates the higher figure. (See Fig. 18.) As Fig. 19 shows, the advantage in creep resistance of steel with coarse grain size holds in this C steel even as low as  $750^{\circ}$ , despite this being far below the generally accepted eutectic temperature. Figure 20 shows a similar behavior in another steel. Coarsening the grain has the disadvantage of reducing the impact resistance; but, as has been pointed out in Chap. 9, Vol. I, if

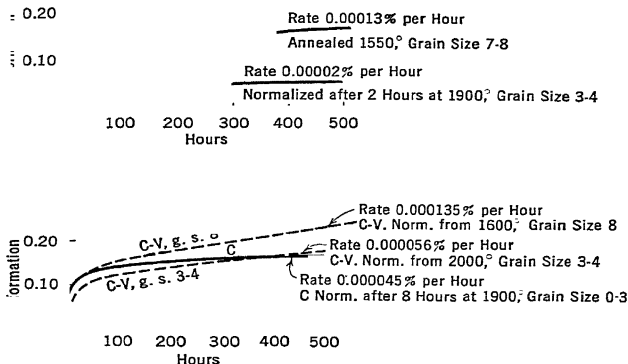


Fig. 18.—Grain size vs. creep in C and CV steels. (Cross and Lowther.<sup>27</sup>)

Upper curves for 0.35 C, 0.55% Mn steel at  $850^{\circ}$  and 7500 lb./in.<sup>2</sup>, in two grain sizes. Lower curves indicate. Morevoer, Figs. 18 and 21 show that a V steel may be decidedly good in creep resistance, even though fine grained. Figure 22 summarizes the behavior of the steels of Fig. 18–21, together with several others that showed analogous behavior.

Results of Jenkins and Co-Workers. Jenkins, Mellor, Tapsell and Johnson<sup>32</sup> studied 5 normalized, killed, plain C steels of 0.13–0.20% C, in which final creep rates at the end of 480 hr. at  $840^{\circ}$  and about 18,000 lb./in.<sup>2</sup> ran, for four of the steels, from 0.0003 to 0.013%/hr., while the

fifth, a "black sheep," showed 0.0108%. They also worked with 7 normalized killed steels of 0.37–0.42% C, whose final creep rates, for the test conditions noted, ranged, on 6 of them, from 0.0004 to 0.0021% per hr., while the seventh, another "black sheep," showed 0.0185%. These "black sheep" were then coarsened by heating to 2010° for 6 hr. and furnace cooled. The higher C steel had its creep rate greatly improved by this treatment, but the creep rate of the low-C steel was actually worse for the coarsened, annealed, state than for the original

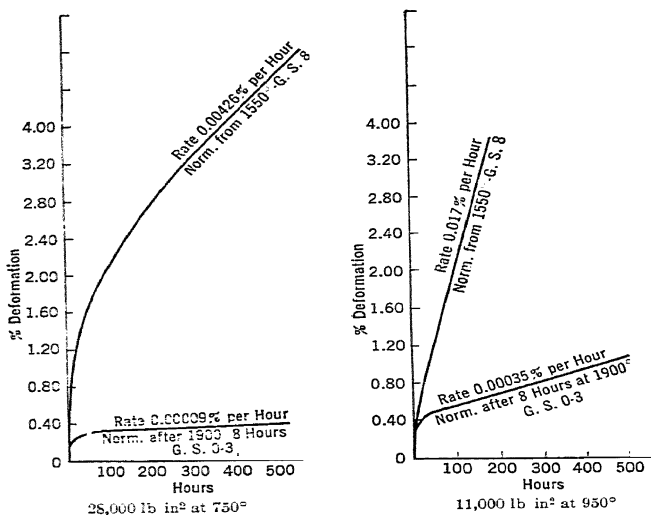


FIG. 19.—Creep, at two loads and temperatures of the same steel, 0.35 C, 0.55% C, Mn, in two grain sizes, produced by normalizing under different conditions. (Cross and Lowther.<sup>27</sup>)

Other curves for this steel, designated "K20," are shown in Figs. 7, 12, 1S and 20.

normalized state. The actual grain sizes are not given nor is information recorded on the coarsening temperatures or coarsening propensities of the steels that might help to clear up the matter.

It has been stated by Clark and White<sup>11</sup> that, in general, creep resistance increases in the following order of melting methods: basic open hearth, arc electric, and induction electric. However, the high-C "black sheep" of Jenkins and co-workers was a basic electric steel.

Weaver<sup>33</sup> has examined a large number of S.A.E. 4330 steels, such as that of Fig. 6, on which he has creep data. At 840°, in terms of

stress to produce a creep rate of 1% in 100,000 hr., he finds that the steels have varied from 10,000 to 40,000 lb./in.<sup>2</sup>. The variation is ascribed to differences in heat treatment, to banded structures, and to differences in grain size, though the grain sizes varied only from around 8 to around 4, and no exceptionally coarse-grained steels whose grain size was definitely known were studied. On scanty evidence, he suggests the possibility of the creep resistance decreasing with further increase in grain size, which is not in accord with the

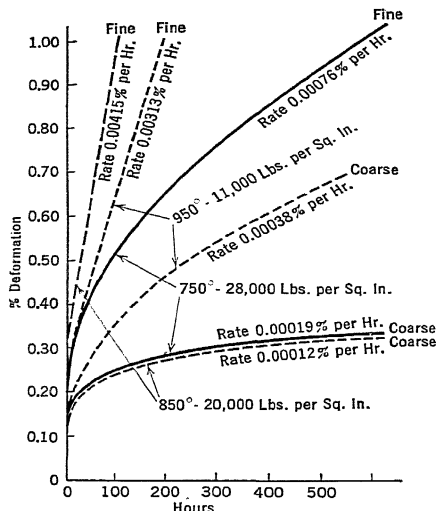


FIG. 20.—Creep behavior of coarse- and fine-grained steels, at loads and temperatures shown. (Cross and Lowther.<sup>27</sup>)

0.51 C, 0.21 Si, 0.75 Mn, 0.10% Al added. Fine, normalized from 1550°, grain size 6-8. Coarse, normalized from 1850°, grain size 2-4. (This is steel No. 56 in Fig. 22.)

work of Cross and Lowther<sup>27</sup> on C steels. However, Weaver concludes that, of the 30,000 lb./in.<sup>2</sup> variation in load-carrying ability of his specimens, 20,000 lb./in.<sup>2</sup> can be ascribed to variation in grain size. Higher temperatures and longer times in heating the steel in the austenite temperature range, producing larger grain size, are favorable to improved creep resistance.

Thus, with the evidence that grain size (true grain size, as established by the actual heat treatment used, which is not necessarily shown by the McQuaid-Ehn test at 1700°) is one of the controlling variables in creep, even at as low as 750° in C steels, an explanation appears for the apparent discrepancies in reported creep tests in which grain size was neglected. By comparing the effect of alloying elements on steels of the same grain size, a clearer picture of the real effects of



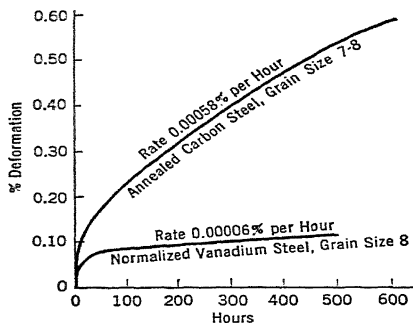


FIG. 21.—Creep behavior of two steels, both fine grained, at 850° and 10,000 lb in.<sup>2</sup> (Cross and Lowther.<sup>27</sup>)

	% C	% Mn	% V	
Carbon.....	0.35	0.55	.....	Annealed at 1550°.
Vanadium.....	0.37	0.70	0.18	Normalized from 1600°.

(See Fig. 18 for behavior of these steels at other loads.)

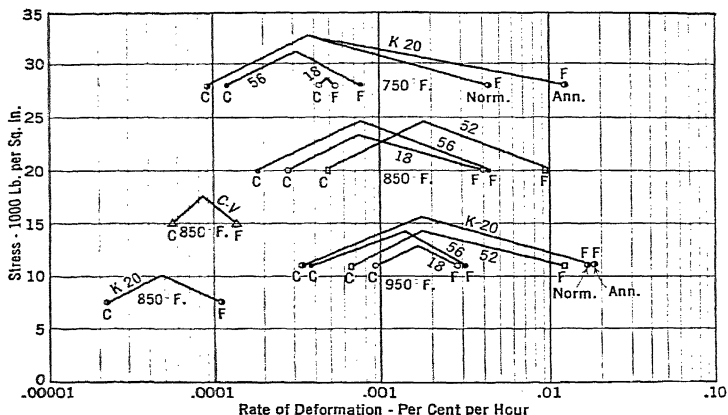


FIG. 22.—Comparison of creep rates, at 500 hr., of five steels at various loads and temperatures, in coarse-grained (C) and fine-grained (F) conditions. (Cross and Lowther.<sup>27</sup>)

In every case, coarsening the grain increases the creep resistance.

Steel No.	Composition, %					Austenitic grain size	
	C	Si	Mn	V	Al*	Coarse	Fine
K20	0.35	0.19	0.55	.....	0.06	0-3	8
52	0.46	0.23	0.54	.....	0.105	2-4	6-8
56	0.51	0.21	0.75	.....	0.10	2-4	6-8
18	0.58	0.14	0.86	.....	0.05	2-4	6-7
CV	0.37	0.16	0.70	0.18	.....	3-4	8

\* Amount added.

the alloying elements will hereafter be obtainable. Such effects have hitherto often been swamped by the factor of grain size.

Obviously, deoxidation practice, methods of grain-size control, normalizing temperature, and the resulting grain size play so large a part in the creep behavior of steel that conclusions as to the effect of alloying elements, drawn on isolated tests, may not be generally valid, although the test correctly represents the behavior of that composition with that melting history and grain size. Since information is lacking in so many respects as to the true relationship of the factors controlling the creep behavior of plain C steels, it is not remarkable that information as to the effect of alloy elements on creep is vague, chaotic, and often apparently contradictory. Only when the effect of an alloying element is so outstanding that it has not been obscured by other effects is the picture clear.

**Effect of Molybdenum.** The beneficial effect of Mo on high-temperature strength is clear. The American Society for Mechanical Engineers<sup>34</sup> has published the following tentative table for permissible design loads for killed plain C and 0.50 (0.40–0.60%) Mo steels.

SUGGESTED ALLOWABLE WORKING STRESSES FOR PLATE STEEL  
OR FORGED FLANGES AND FITTINGS

Temp. Deg.	LOAD—LB./IN. <sup>2</sup>		CASTINGS	
	C	C-Mo	C	C-Mo
650	14,000	14,000	10,000	10,000
700	13,300	14,000	9,500	10,000
750	11,900	14,000	8,500	10,000
800	10,000	14,000	7,200	10,000
850	7,850	12,000	5,800	8,800
900	5,600	10,000	4,400	7,800
950	3,800	8,000	3,200	6,700
1000	2,000	5,000	.....*	4,600

\* Limited to 950°.

All the steels are of a grade to give 70,000 room temperature tensile. The C steels are described in the A.S.T.M. specifications A-149-36; A-150-36 Grade B; A27-36T Grade B and B<sub>2</sub>, and A95-36. The C-Mo steels in A.S.T.M. A204-37T Grade B, A95-36, and A157-36C<sub>1</sub>. Nickel steel of 2.00–2.75% Ni, 203-37T, is classed with the C steels as to allowable stresses.

The working stresses listed are not creep-test values, but are derived from creep tests with a factor of safety and checked by experience.

**Creep of Carbon-Molybdenum Steel.** Creep tests given in advertising literature of the Babcock and Wilcox Tube Co.<sup>35</sup> and of the Timken Steel and Tube Co., comparing plain C and C-Mo steels, are shown in Fig. 23. Compare also Fig. 4, Chap. 9. That 0.50% Mo has nearly as much effect as 1% is also shown in this figure. Bailey<sup>36</sup> remarks that 0.50% Mo produces about  $\frac{3}{4}$  as much effect on creep resistance as does 1%.

Miller, Campbell, Aborn, and Wright<sup>36</sup> made creep tests on C-Mo steel of 0.11 C, 0.17 Si, 0.47 Mn, 0.54% Mo that had been normalized from 1650° (grain size 6-7) and drawn at various times and temperatures. The creep tests were run about 3000 hr. at 1100°. Depending

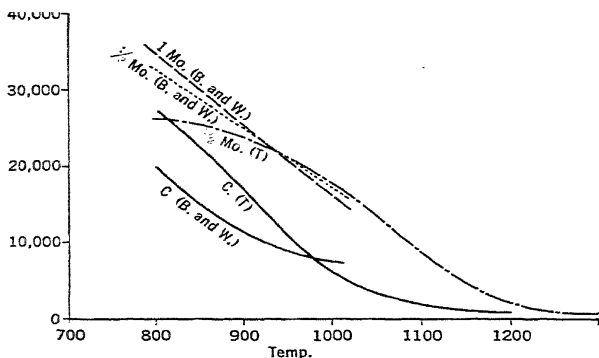


FIG. 23.—Creep properties of C and C-Mo steels. 1% in 10,000 hr. (After advertising literature of Babcock and Wilcox Tube Co., and Timken Roller Bearing Co.)

on the time and temperature of tempering, a creep rate of 1% in 10,000 hr. at loads of 2300 to 3500 lb. in.<sup>2</sup> was produced. (See Fig. 24.) The initial and final structures were compared. An initial structure with a fine precipitate, apparently of Mo-rich carbide, throughout the ferrite ground mass, produced by tempering 5 hr. at 1300° was found to give the highest load-carrying ability. This treatment did not decrease the impact resistance from the 50 ft.-lb. of the normalized, untempered, state. The poorest results were on material tempered 168 hr. at 1300°, which treatment had caused this precipitate to agglomerate and coarsen and had dropped the impact to about 30 ft.-lb. The precipitate, in the dispersion produced by tempering 5 hr. at 1300°, was structurally stable in 3000 hr. under load at 1100°. The benefit of Mo is ascribed to the strengthening effect of this precipitate and to its stability. This stability of the Mo-containing carbide is presumably the

reason for the beneficial effect of Mo on temper brittleness, which is thought to arise from carbide solubility and precipitation. Jenkins and co-workers<sup>32</sup> also made creep studies on steels of 0.50 and 1% Mo in various conditions of heat treatment. Coarsening was found advantageous, spheroidizing disadvantageous, in tests at 1020°. Clark and White's<sup>11</sup> curves for 0.50% Mo steel are shown in Fig. 8.

**Tungsten.** Plain C-W steels find little use, although the effect of W is very similar to the effect of Mo, because it takes at least 1% W to confer equivalent creep resistance to that conferred by 0.50% Mo. Figure 25 indicates, indeed, that it will require more than 1% to equal

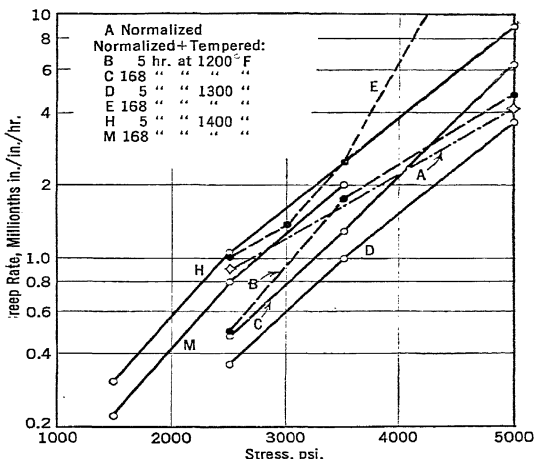


Fig. 24.—Influence of heat treatment on creep of C-Mo steel at 1100°.

Composition, %			
C	Si	Mn	Mo
0.11	0.17	0.47	0.54

Semi-log plot. (Miller and co-workers.<sup>26</sup>)

0.50% Mo with 5% Cr. In combinations with other elements some producers prefer the W addition because of free-sealing behavior in forging. In respect to properties of the steel there is no strong evidence to lead to the conclusion that W makes a better high-temperature steel than Mo; hence, on account of cost, Mo has received most of the attention. Most users would probably allow substitution of a suitably larger amount of W for Mo. Steel Q in Fig. 11 is a W steel.

**Vanadium.** Vanadium has not yet had the attention it deserves in creep-resistant steels. Figure 18 and 21 show that even a fine-grained V steel is superior in creep resistance to a fine-grained Al-

treated C steel of similar grain size. The use of V may therefore favor the combination of good creep resistance plus good impact.

Bailey<sup>10</sup> has made high load tests at 1020° and 24,700 lb. in.<sup>2</sup> on steels of 0.20 C, 0.50% Mo, without V, and with additions of 0.20 and 0.40% V. Under these conditions the C-Mo stretched 2% in 200 hr., while the steels to which V was added stretched 1.5% in 1500 hr. The rates of creep increased on all three steels soon after the times stated. The evidence did not reveal how the steels would act at reasonable loads, but it gives an interesting hint as to probable usefulness of V. Since V is known to produce secondary hardening by a precipitation effect, it would be necessary to establish that the steel is stable against overaging so that the strength will be permanent.

Clark and White<sup>11</sup> cite an air-hardening steel of 0.45 C, 0.77 Si, 0.47 Mn, 1.40 Cr, 0.9S W, 0.25% V, normalized from 1725°, drawn at 1300°, grain size 8, which they suggest for use at 1000–1100° as spheroidization occurs at 1200°. Their creep values are:

	1% in 100,000 Hr.	1% in 10,000 Hr.
800°.....	48,000	68,000
1000°.....	13,750	24,000
1200°.....	540	1,200

**Phosphorus Creep-resistant Steels.** A possible substitute in low-C steels, for some or all the Mo is P, since Cross and Krause<sup>37</sup> found that, of two normalized steels, each with 0.10 C and 1% Cr, one with 0.20% Mo, after 1000 hr. at 550° and a load of 12,000 lb. in.<sup>2</sup>, had stretched 0.088%, while one with 0.20% P had stretched 0.089%. At 950°, under the same load, a 0.10 C, 1 Cr, 0.50% Mo steel stretched 0.091% while one of 0.10 C, 1 Cr, 0.50% P stretched 0.112%. These values are sufficiently of the same order to indicate that P steels may be worth study for creep resistance. The 0.50% P steel was brittle, but the one with 0.20% P was tough after creep (av. 64 ft.-lb. Izod). Hence replacement of part of the Mo was tried, a 0.17 C, 0.92 Cr, 0.20 Mo, 0.17% P normalized and drawn steel being run at 950°, 12,000 lb. in.<sup>2</sup>. This steel stretched 0.116% in 1000 hr., and averaged 43 ft.-lb. Izod after creep.

**Manganese.** Little definite evidence is available as to the effect upon creep of increasing Mn in a C steel. Qualitatively, it would appear that raising Mn to around 1.25% may be slightly helpful, but it might be that this is primarily due to an increased tendency toward grain coarsening. White and Clark<sup>17</sup> have presented data (Fig. 9) for

steel of 1.25 Mn, 0.25% Mo, which may be compared with their data for 0.50 Mn, 0.50% Mo (Fig. 8).

**Chromium.** Chromium, by itself, has a slight strengthening effect against creep, but as Fig. 25 shows, even 5% Cr, without Mo, is not highly creep resistant. Bailey<sup>6</sup> points out that an addition of 1% Cr to 0.50% Mo steel strengthens under some conditions and weakens under others, as shown in Fig. 26. This is characteristic of pearlitic steels; a comparison between two steels at one load and temperature may not be at all valid under another set of conditions. Fleischmann<sup>38</sup> shows, Fig. 29, the effect of Cr in steels with 0.50% Mo.

**Effects of Chromium on Resistance to Oxidation.** Where corrosion and oxidation resistance are required, steels, for use at the upper end of the temperature range in which the steels are pearlitic, are usually based on Cr. When in the order of 1 or 2% Cr, especially with other alloys, the use of Cr gives somewhat higher load-carrying ability than in its absence, but further increase in Cr seems to decrease the load-carrying ability. Thus the 12–14% plain Cr stainless type, without other alloy additions, is not of high load-carrying ability. Increase in Cr does steadily increase resistance to oxidation, and to many of the types of corrosion encountered in high-temperature service.

**Ferrite Strengtheners—Nickel, Copper, Silicon.** By themselves, the elements Ni, Cu, and Si, which strengthen ferrite at ordinary temperature, are almost without effect on creep at high temperature. Nickel is used in some bolting steels that are quenched and tempered, for relatively low-temperature use, but in steels for high-temperature creep resistance its effect is neutral, save that it has a tendency to reduce scaling.

Copper, in steels with Mo, has been reported as helpful by some observers, but such action is not well substantiated.

Silicon seems to have no strengthening effect by itself, but is used to confer increased corrosion and oxidation resistance. Silicon-Mo steel has been suggested by Fleming<sup>39</sup> for high-temperature service. Such a steel is indicated by Fig. 29 to act about the same as in the absence of Si. Silicon appears in several complex steels.

**Chromium-Molybdenum Steels.** Chromium steels varying from about 2 to about 9% Cr, with 0.50% or more Mo, find wide use as still tubing in the petroleum industry. The 5 Cr, 0.50% Mo composition is the most common composition, but for higher or lower temperatures of service, and for use with oils of varying corrosiveness, the compositions on each side are used. If the oil is not corrosive, and the temperature not so high that oxidation of the outside of the tube is feared, the Cr is omitted and plain C-Mo tubing used.

The benefit of Mo in preventing, or postponing for a long period, the embrittlement that may occur in its absence has already been mentioned. Figures 25-29 show data on steels of varying Cr and Mo

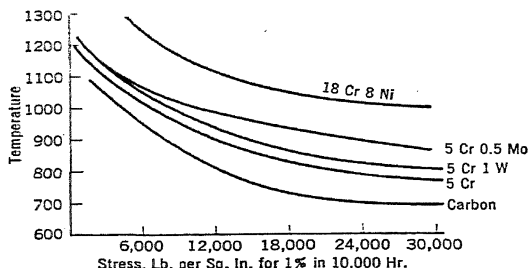


FIG. 25.—Creep of steels for oil refinery tubes. (Wright.<sup>36</sup>)

content. Babcock and Wilcox data in Fig. 28 show that 2 Cr, 0.50% Mo is slightly better than 5 Cr, 0.50% Mo, while 9 Cr, 1.50% Mo is much stronger to about 1050° but thereafter less strong than the 2 Cr, 0.50% Mo.

It is stated<sup>72</sup> that W does not remove brittleness of 5% Cr steel in the cold, whereas Mo does, so that the latter is usually preferred.

Figure 25 allows comparison with 5% Cr free from Mo and also shows how much superior to these pearlitic steels is the austenitic 18 : 8. Data from Turner, quoted by Mochel<sup>40</sup> for steel of 0.10 C, 13.5 Cr, 3% W, would fall midway between the curves of Fig. 25 for 5 Cr, 0.5% Mo and 18 : 8, while data from Norton<sup>19</sup> for plain 13% Cr, without Mo or W, would fall close to the curve of Fig. 24 for 5% Cr without additions.

**Silicon in Chromium-Molybdenum Steel.** Newell<sup>41</sup> studied the effect of Si in steels of 2.00 Cr, 0.50% Mo and 5.00 Cr, 0.50% Mo and concluded that, although Si improved the resistance to scaling, it decreased the impact resistance, and that, though creep resistance at 800° might be somewhat improved, it was injured at 1000°. He indicated that only in resistance to scaling is anything important gained by raising Si above the amount needed for thorough deoxidation of the melt. An effect of Si in a lower Cr steel, that would not be predicted is indicated by the Timken data in Fig. 29 in that so much of a hump appears in the 900-1050° range at the particular combination of Cr, Si, and Mo used in the 11½ Cr-Mo steel, whereas the curves for 21½ Cr-Mo, and Si-Mo are essentially identical.

Data by Miller, Campbell, Aborn, and Wright<sup>38</sup> on a steel of lower

Si, i.e., of 0.20 C, 0.49 Si, 0.49 Mn, 1.56 Cr, 0.65% Mo, normalized and drawn, give the points shown in the plot, which would give a curve without the hump.

**Aluminum and Silicon in Chromium-Molybdenum Steels.** Increasing the Si in a 0.15 C, 0.35 Mn, 1.00 Mo, 0.70% V steel from 0.30 to 1% is alleged by "Kohle und Eisenforschung"<sup>42</sup> to raise the creep resistance at 930° from 31,000 to 54,000 lb./in.<sup>2</sup> and at 1020° from 17,000

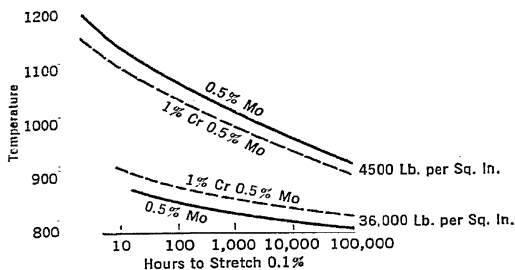


FIG. 26.—Note reversal in position with change in temperature of test. (Bailey.<sup>6</sup>)

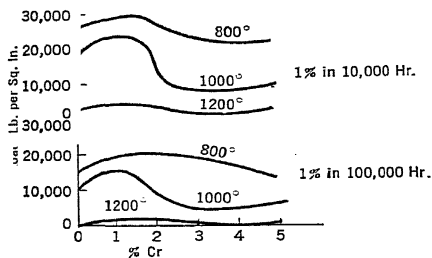


FIG. 27.—Effect of Cr in annealed steels carrying 0.50% Mo. (Fleischmann.<sup>35</sup>)

FIGS. 26 and 27.—Effect of Cr in 0.50% Mo steel.

to 36,000 lb./in.<sup>2</sup>. Addition of 1.5% Cr is said not to affect the creep of these steels. The criterion of creep is not stated, so the figures are valueless for comparison with other data.

Dauvergne<sup>43</sup> adds 1% Al and 1% Si to the 5 Cr, 0.50% Mo steel and claims that this improves the high-temperature properties. He also claims improvement over a plain C, 0.50% Mo steel by addition to it of 0.25 Cr, and 0.25% Ni, and likewise favors a steel of 0.15 C, 0.10 Si, 0.80 Mn, 0.50 Mo, 0.50 Cr, 0.50% Al, this Al addition being preferred to calorizing.

Fleischmann and Williams<sup>44</sup> discuss a family of steels for tubes in



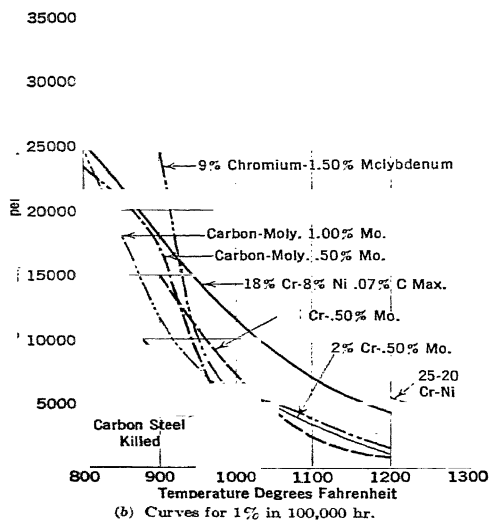
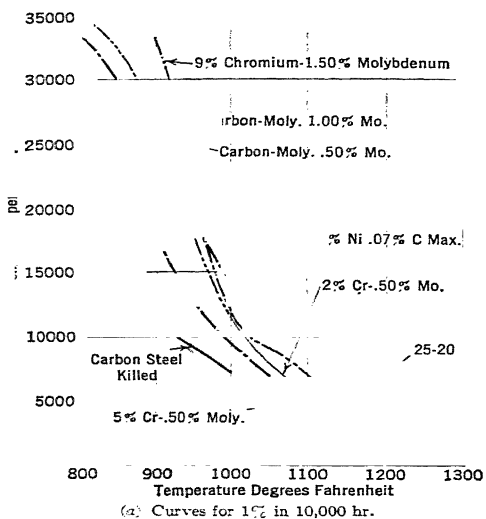
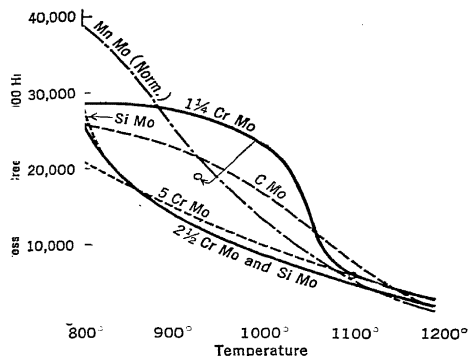


FIG. 28.—Creep properties of tube steels. (Babcock and Wilcox Tube Co.<sup>91</sup>)

refining sour, i.e., corrosive, crude oils. These contain 0.15 C max., 0.45–0.65 Mn, 0.30 Mo max., 1.00–2.00 Si, 0.60–0.80% Al, and Cr in one of the following ranges: 0.75–1.25, 1.75–2.25, 2.75–3.25, or 4.00–6.00%.\* Such a steel with 3% Cr is considered equivalent to one of 5 Cr, 0.50% Mo without the Si and Al, and the 5% Cr steel with Mo, Si,



\* For comparisons of resistance to oxidation of various CrSiAl steels, see White, Clark and McCollam<sup>94</sup>

FIG. 29.—Creep properties of steels containing Mo. (Timken Steel and Tube Co.<sup>90</sup>)

Steel	Composition, %					
		Si	Mn	Cr	Mo	
1 1/4 Cr Mo	0.15 max.	0.50–1.00	0.30–0.60	1.00–1.50	0.40–0.60	
2 1/4 Cr Mo	0.15 max.	0.50–1.00	0.50 max.	2.25–2.75	0.40–0.60	
5 Cr Mo	0.15 max.	1.15–1.65	0.30 max.	.....	0.40–0.60	
C Mo	0.10–0.20	0.25 max.	0.30–0.60	.....	0.45–0.65	
5 Cr Mo	0.15 max.	0.50 max.	0.50 max.	4.00–6.00	0.45–0.65	
Mn Mo	0.15 max.	0.15–0.30	1.10–1.40	.....	0.20–0.30	

Annealed at 1550°

Normalized from 1725° and drawn at 1200°.

The point indicated by the arrow connected to the 1 1/4 Cr Mo curve is for a steel of 0.20 C, 0.49 Si, 0.49 Mn, 1.56 Cr, 0.65 % Mo, normalized from 1750°, drawn at 1375°, as shown by Aborn and co-workers.<sup>36</sup> Another point for this steel and treatment, from Aborn's test at 1100°, lies on the Timken 1 1/4 Cr Mo curve.

and Al is stated to be almost as good as 18 : 8. However, Gottignies<sup>45</sup> said that in a 5% Cr steel with 0.50% Mo, the addition of 0.50% Al injured the creep resistance. The evidence is thus highly contradictory. Perhaps each individual comparison is correct for the particular steels compared and some unrecognized variable is acting.

**Effect of Titanium.** Bailey<sup>46</sup> reports that addition of Ti to a C-Mo steel reduces its creep resistance. In such a steel the C would be quite strongly bound to Ti as carbide, so the Mo carbide might not be able to form. Bailey also stated that steel of 1.00 Cr, 0.50% Mo was inferior in creep to one of that Mo content without Cr. Fleischmann and Williams<sup>44</sup> add some 0.70% Al to such steels and find that it reduces air hardening, much as Ti does.

**Chromium-Titanium.** Comstock and Clark<sup>47</sup> compared the high-temperature properties of steels of (A) 0.22 C, 0.38 Si, 0.57 Mn, 17.85% Cr, and (B) 0.30 C, 0.91 Si, 0.70 Mn, 17.75 Cr, and 0.25% Ti. The ratio of Ti to C was too low to "fix" the C as is done with larger

additions for stabilization. The steels were hard as forged, *A* 420, *B* 460 Brinell, and were annealed at 1650° to 165 and 180 Brinell. The room temperature properties of annealed specimens were:

	Tensile	Yield (0.1% Deformation)	Elong. %	R.A. %
<i>A</i> .....	80,500	42,000	29	53
<i>B</i> .....	90,500	52,500	24	47

Short-time tensile tests showed *B* to be the stronger at 1000°, 55,000 tensile against 43,500 for *A*; but at 1200 and 1400° there was little difference. Creep tests at 1100° under 6000 lb. in.<sup>2</sup> for 1000 hr. gave a final creep rate of 2.95% in 10,000 hr. for *A* and 2.65 for *B*, not a large difference. The specimens, deformed in the creep test, were broken in tension at room temperature and showed their original strength. The ductility of *A* was reduced, to 19% elong., 25% R. A., while that of *B* was unaffected. Similar tests were made on analogous 0.20 C, 17.50% Cr steels with (*C*) 1.10% and (*D*) 2% Ti. These steels were soft as forged, about 135 Brinell, and maintained about this hardness after annealing at 1650 and 1550°, respectively. They showed, at room temperature, annealed:

	Tensile	Yield (0.1%)	Elong. %	R.A. %
<i>C</i> .	75,500	44,500	26.5	44
<i>D</i> .	72,000	42,500	23	41.5

In the short-time, high-temperature tests, both steels were weaker than *A* at the test temperatures of 1000, 1200 and 1400°. In creep at 1100° Steel *C* was stretching at the rate of 40% in 10,000 hr. after 500 hr. and *D* at 6.8% after 1000 hr. *C* had been strained nearly to failure; *D* showed no reduction in ductility after creep.

Although high Ti was not helpful in producing high-temperature strength in these annealed steels, the addition of 0.25% Ti seemed useful in avoiding loss of ductility.

**Heat-resistant Steels or Alloys.** The completely pearlitic steels, even the highly alloyed ones, are seldom applicable for temperatures above 1200°. Those low in Cr are insufficiently resistant to oxidation, and the very high, plain Cr steels, e.g., 30–35%, are not strong. Such a composition as 9 Cr, 1.50% Mo gives a useful combination of properties, but as Fig. 28 shows, even this has little load-carrying ability above 1200°. Fleischmann<sup>38</sup> gives the following as suitable for oil-refinery equipment up to the temperatures shown: killed C steel, 900°:

C-Mo, 1000°; C-Mo plus 1.25% Cr or 1.5% Si, 1100°; 2 to 5% Cr, with Mo, 1200°; 5 Cr, 1.5% Si, with Mo, 1300°; 18 Cr, 8 Ni, 1300°, with 16 Cr, 13 Ni, 3% Mo for higher temperatures.

For the range above 1200 up to 2100°, choice is practically confined to steels that are initially austenitic, so that in heating and cooling they will not pass through a transformation range. To resist oxidation, the steels are high in Cr, and the Cr is accompanied by enough of some other element or elements to make the steel completely austenitic, or at least predominantly so. The usual addition is one of at least 7 to 8% Ni, since that is the least amount that will make an 18% Cr steel austenitic. The lower end of the 1200–2100° range is therefore most commonly served by 18 : 8, with or without stabilizing or strengthening additions. For higher temperature service both the Cr and Ni may be raised, 26 Cr, 12% Ni being a very common composition. For example, it is used in radiant tubes. As Fig. 28 shows, that composition has a little higher load-carrying ability than 18 : 8. Chromium up to about 30%, with 8 to 12% Ni, is used in this general type. Chromium around 15%, with Ni at 35% in one grade and 65% in another, is also popular. Inasmuch as these compositions are so highly alloyed, they rank as alloys rather than as steels.

No heat treatment, in the ordinary sense, is possible in these austenitic alloys. To stabilize the materials against harmful carbide separation of the precipitation-hardening type, they are usually given a final heat treatment from above the carbide solution temperature, and either air cooled or quenched in order to hold the carbides in solution. The problem of carbide separation in 18 : 8, to be discussed in the next chapter, is typical of similar problems with all the alloys.

Some improvement in load-carrying ability is generally considered to be made by addition of up to about 4 or 5% W or 2 to 3% Mo, though there is less certainty as to the degree of benefit, the economic amount to add, or the relative effects of Mo and W than is the case in pearlitic steels.

Hatfield <sup>48</sup> recommends the following for service up to about 2000°, and considers the first two the strongest steels available at such temperature, even though at 1830° proper design values are below 100 lb. in.<sup>2</sup>

C %	Si %	Mn %	Ni %	Cr %	W %
0.23	1.65	0.40	12	23	3
0.35	1.70	0.65	7.25	19	4
0.12	1.50	0.35	21	24	....

The load-carrying ability of heat-resistant alloys at 1800–2000° has sometimes been considerably overstated. Pilling and Worthington<sup>14</sup> in 1931 cited the estimates of various designers for the permissible fiber stress of 0.50 C, 35 Ni, 17% Cr castings, which ran from 290 to 600 lb. in.<sup>2</sup> at 1800° and 140 to 300 lb. in.<sup>2</sup> at 2000°, save for one set of figures, obviously extrapolated, which figured 1250 lb. in.<sup>2</sup> for both temperatures.

In order that the metal may flow readily in casting, the C in castings is often in the range 0.35–0.50% so that the carbide embrittlement problem cannot be avoided by the expedient of keeping the C low, as may be done in wrought steels.

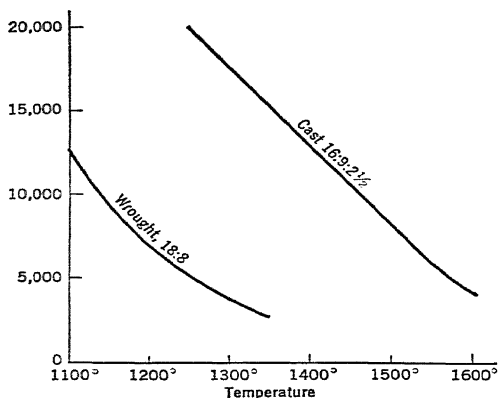


FIG. 30.—Creep values reported for extension at the rate of 1% in 10,000 hr.

Composition, %			
C	Cr	Ni	Mo
0.07	18	8	
0.20	16	9	2.5
wrought (Norton, <sup>19</sup> Cross. <sup>2</sup> )			
cast (Kanter, <sup>51</sup> )			

Tucker and Sinclair<sup>49</sup> found that at 1600° (using cast specimens not provided with shrink heads) an alloy of approximately  $\frac{1}{3}$  each of Fe, Ni, and Cr showed superior creep resistance. Rosenhain and Jenkins,<sup>50</sup> using "accelerated" methods, concluded that a 30 Cr, 30 Ni, 4% W alloy had exceptional properties, and the Crane Company<sup>51</sup> is quoted for data on 18 : 8 plus 2.5% Mo, which are shown in Fig. 30, compared with 18 : 8. Such information indicates that Mo and W may be helpful in austenitic steel, though not all authorities agree in this. Newell<sup>52</sup> remarks that 18 : 8 with 3% Mo contains delta iron, i.e., is not fully austenitic, whereas 16 Cr, 13 Ni, 3% Mo is fully austenitic and preliminary creep tests upon it indicate it to be more creep resis-

tant at high temperatures than 18 : 8. Fleischmann<sup>38</sup> also rates this as having superior creep properties. Rohn,<sup>53</sup> by accelerated methods, concludes that an alloy with 60 Ni, 15 Cr, 7 Mo, 2% Mn has the best load-carrying ability at very high temperatures of the heat-resistant alloys. Tapsell<sup>2</sup> states that addition of Mo and W is most effective in austenitic steels. Krivobok<sup>54</sup> states that W is the only element notably increasing short time and creep resistance of 12-14 and 16-18% plain Cr steels, though boron shows some promise.

Brown<sup>55</sup> suggests a steel of the 18 : 8 class, for use as castings, with Cr 20 min., Ni 8 min., C 1.20 max., and Al ranging from 0.50 to 5%, which steel he credits with a very high safe load at 1800°. One might expect variations in properties in material of so wide an analysis, especially in view of the extremely high C stated as permissible.

Hatfield<sup>56</sup> finds that addition of 1.50 Al and 0.90% Ti to 18 : 8, which addition makes precipitation hardening possible, greatly improves creep resistance. Oertel and Schepers<sup>57</sup> favor the addition of 2.0-2.5 Si and 0.5-1% Al to 18 : 8, and Russian reports are in agreement with the claim that both strength and resistance to oxidation are improved by the additions. In the literature of heat-resisting alloys, an extensive annotated bibliography by Hatfield<sup>48</sup> deserves mention.

When the austenitic Ni-Cr steels are used for high-temperature service in contact with gases containing SO<sub>2</sub>, it is customary to hold the Cr high, and lower the Ni, because Cr is resistant to, and Ni attacked by, this gas. A sufficient preponderance of Cr may allow the use of appreciable amounts of Ni.

**Chromium-Manganese Alloys.** Comment should be made that Cr-Mn-Fe alloys, which, it will be noted in the next chapter, are competitors to the austenitic Ni-Cr-Fe steels and alloys for some corrosion-resistant uses, are also potential competitors in heat-resistant service. The amount of Cr required for resistance to oxidation will be about the same whether it is accompanied by Ni or by Mn. Roughly, twice as much Mn as Ni is required to make a Cr steel austenitic. That is, the analog of 18 Cr, 8 Ni is around 18 Cr, 18 Mn. A steel of 18 Cr, 8 Mn is not fully austenitic. Such a composition, however, plus W and Ti, i.e., 0.40 C, 0.40 Si, 8.90 Mn, 17.00 Cr, 2.20 W, 0.28% Ti, is alleged by Wolfson and Borzdika<sup>58</sup> to be superior in creep resistance to 18 Cr, 8 Ni; but since their comparison is on the basis of tests of less than 150 hr. duration, little credence can yet be placed in the claim without further corroboration.

Becket<sup>59</sup> studied the addition to steel of around 0.10 C, 18 Cr, 9% Mn of 3% W and of 2.5 and 5% Co in respect to short-time, high-temperature strength. At 1100° the tensiles were, no addition, 40,000;

W. 46,000; 2.5 Co. 47,000; 5 Co. 50,000; and at 1650°. 9,000, 10,500, 14,500, and 19,000, respectively. The 5% Co steel was higher in C than the others; it had 0.18%.

Mony penny<sup>60</sup> refers to a Krupp steel of 0.50 C, 13 Cr, 5 Mn, and 5% W. Schmidt and Legat<sup>61</sup> studied some of the high-temperature properties of steels with 0.14–0.21 C, with 19 Cr, 8 Mn, 1 Si, 1% Ni; 17 Cr, 15.5 Mn, normal Si and 1.35% Ni; and 14 Cr, 16 Mn without Si or Ni. A decrease in impact toughness when annealed at 950° was noted for the first and at still higher temperatures for the second. The first would have a duplex structure containing both ferrite and austenite. The second would be largely austenitic. No comparable data were given for the third, which would be essentially austenitic. Monypenny, in discussing this work, questions whether the loss of toughness is any greater than in the usual Cr-Ni heat-resistant alloys used for castings. Schmidt and Legat made a few so-called creep tests at high loads and for very short times that are valueless in themselves, but which led them to conclude that the austenitic 14 Cr, 16% Mn steel would have useful creep-resisting properties while the 17 Cr, 15.5 Mn, 1.35% Ni would be less useful, a discrepancy that is not readily explained. However, Figs. 7 and S, Chap. 20, show that these compositions are close to the dividing line between the austenite and austenite plus ferrite fields, and small changes may throw the alloy to one side or the other.

Figures 7 and S, Chap. 20, show that the curvature of the dividing line and the necessity of avoiding compositions that give the brittle sigma phase, limit the Cr and Mn content more than in the austenitic Cr-Ni steels. That is, the fully austenitic steels with Mn cannot contain more than about 15% Cr, which is not enough for oxidation resistance at very high temperatures. So the combination of high Cr for resistance to scaling and the fully austenitic structure required for best creep resistance cannot be obtained in the Cr-Mn series as it is in the Cr-Ni series. However, the Cr-Mn series will undoubtedly meet some high-temperature needs, as well as a portion of the requirements for corrosion resistance.

**Steels for Use Against Hydrogen at High Temperatures and Pressures.** In steels for use in the high-temperature, high-pressure synthesis of ammonia from  $H_2$  and  $N_2$ , carbide stabilizers are used.\* These are Cr, V, Mo, and W.  $Fe_3C$  is readily decarburized by H and ordinary steel is embrittled in such service; but with enough of these carbide-formers, as has been shown by Vanick and co-workers,<sup>62</sup> Kesting,<sup>63</sup> and Maxwell,<sup>64</sup> embrittlement is avoided. Maxwell exposed

\* Naumann<sup>66</sup> advocates Ti as carbide stabilizer in addition to Mo and Cr.

specimens for 1500 hr. at 1000 atmospheres and 930°. S.A.E. 4130, 6120, and 7260 came through this test unscathed in tensile strength or ductility, whereas Fe, 1012, 1045, 2315, 3115, and 5120 had lost all ductility and been halved in strength in half that time. Ranque<sup>65</sup> advocates a steel of 4.5 Cr, 0.65 Mo, and 0.70% V for this purpose.

**Automotive Valve Steels.** A difficult high-temperature service in which must be combined strength, hardness, and resistance to corrosion by products of combustion, which may carry the decomposition products of tetra ethyl lead, is that of automobile and aircraft engine exhaust valves.

Chromium is the element chiefly relied upon for these types of resistances. For moderate duty an air-hardening ferritic steel of about 0.50 C, 9 Cr, 3% Si is employed. This does not retain its hardness at elevated temperatures as well as the austenitic steels, and more Cr improves the resistance to scaling, so highly alloyed steels of the Ni-Cr class, with Si or W or both, are in use. Some of the type compositions used are:

% C	% Cr	% Ni	% Si	% W	% Mn	% Mo
0.40	19.50	8.0	3.25	....	....	....
0.50	14.00	14.25	0.50	2.25	....	....
0.60	15.50	14.25	2.75	2.25	....	....
0.25	21.50	11.50	1.00	....	1.00	....
0.45	25.00	3.00	1.00	....	1.30	3.00
0.55	26.00	5.00	0.40	....	0.35	7.00

In general, these are not heat treatable in the ordinary sense, though, as with 18 : 8, a stabilizing heat treatment may be given to minimize carbide precipitation in service.

Other austenitic steels on the high side in Cr and with the other alloying elements balanced so as to produce greater hardness at the most elevated temperatures of operation, combined with forgeability, are under development. Some of these, in experimental valve use, have considerably out-performed steels of the above list, so the range of compositions likely to be utilized in the future, for extremely severe duty, will ultimately be extended.

**Seizure.** A troublesome problem appears when two parts are in rubbing contact at high temperature, in the absence of a lubricating film. For example, valves and valve stems operating in superheated steam may gall or seize, that is, incipient welding takes place. If the coating parts are adjusted by heat treatment to show a considerable difference in hardness, seizure may be lessened, but wear may be in-



creased. Nitrided steel is notable in the variety of other materials against which it will operate without seizure. Few general rules for avoidance of seizure can be laid down other than that some film (such as one of lead from a leaded bearing metal, graphite from cast iron, or one of non-metallic nature, such as a nitrided skin, or an oxide coating) between the coating parts may be helpful. The topic has been discussed by Mochel<sup>64</sup> and by Bolton.<sup>67</sup>

**Requirements for Bolting Stock.** Turning from the consideration of the higher temperature ranges to one in which  $1000^{\circ}$  would be considered as a rather high operating temperature, we find that the materials chosen, and the heat treatment applied, shift accordingly. The problem of holding flanges together tightly enough to prevent leakage on pipe and fittings for superheated steam, in power plant service, or for hot oil or the like, is a special one, since the bolt is initially tightened so that it is under rather high elastic stress. When the bolt is heated and creep occurs, the tension decreases, i.e., the bolt relaxes. Leakage or the necessity for frequent tightening may then ensue if the creep resistance of the bolt is insufficient.

Even though the bolts and flanges may be covered with insulation, the bolt temperature is seldom extremely high. This fact allows, and the necessity for high yield strength when the bolt is tightened practically demands, that steels be used in the quenched and tempered condition, in contradistinction to the general practice of using normalized and drawn steels for other high-temperature services.

**Evaluation of Load-carrying Ability by Variable Load or Temperature Methods.** Some evaluation of the creep resistance of bolting stock may be made by regular creep tests at one load and temperature, but such a test does not evaluate the ability of the steel to strain harden under the high initial stress of the original tightening. For a more direct evaluation of this property, one may, as Robinson<sup>68</sup> does, start a creep test at a high load, allow the specimen to elongate a fixed amount, and when that deformation has been reached, lower the load by a predetermined amount, a selected deformation to be reached again, and so on. This is a "down-step" method of testing. The difficulty is in selecting the program of loads and extensions so that different steels may be clearly compared; hence the method finds little use except in specific relation to bolting stock. Methods have been tried in which, after the specimen under fixed load has elongated a given amount, the temperature is automatically reduced until elongation ceases, the final temperature being the criterion for comparison.

"Up-step" methods, in which a low load is first applied for a definite period and the load then raised for another definite period and so on,

have been suggested, but Kanter and Spring,<sup>9</sup> White and Clark,<sup>13</sup> Tapsell<sup>2</sup> and others have shown that results from none of the modified methods can be directly compared with those of the usual constant-load, constant-temperature creep tests. Many of these methods have been tried out in the hope of finding a rapid substitute for the regular creep test, without realization of the hope. Uncertainties as to the effect upon the steel of application of a different load or temperature from the final one make the interpretation of such results very difficult. Of all the variable load or variable temperature methods that have been suggested only the relaxation test, and that only in respect to bolting stock, has much interest outside pure research into the mechanism of creep.

**Relaxation.** A direct method of determining the relaxation properties of bolting stock that comes close to duplicating the service conditions in bolts, has been described by Mochel.<sup>69</sup> A long bolt of the material to be tested is tightened upon a heavy, rigid frame of austenitic 18 : 8 steel or heat-resisting alloy by screwing up the nuts at the ends until an extensometer applied to gage marks on the bolt indicates the desired initial tension as calculated from the modulus of elasticity. Bolt, frame, and all are placed in a furnace and held at the desired test temperature for 500, 1000 hr., or more. The assembly is then removed, cooled to room temperature, and the residual stress measured with the extensometer applied to gage marks. Insufficient data are on record to allow appraisal of bolting stock by this method, but it is of interest to note that in a comparison of coarse- and fine-grained C steel by this method,<sup>27</sup> at 850°, the coarse-grained steel showed superiority just as it did in the standard creep tests. Available data on high-temperature properties of bolting stock are mostly those obtained by the regular constant-load, constant-temperature creep method.

When bolting stock for high temperature use was not required to operate above 750°, S.A.E. 3140 Ni-Cr steel, quenched and tempered, was commonly used. Since temper brittleness is likely to occur in such steel without Mo, and long sojourn at elevated temperature may have an analogous embrittling effect, 3140 has been replaced, in higher temperature service, by steels of higher creep resistance. Kanter<sup>51</sup> points out that, though 3140 does lose impact resistance, especially at 900°, it retains sufficient toughness in contrast to the brittle behavior of Bessemer C steel, which was used in the early days and at much lower temperatures, so the choice of other steels has been made primarily on the score of creep resistance.

The A.S.T.M. specifications for bolting stock, and for comparison, those for piping, for service at 750–1100°, are summarized as follows:

## A 193-37 T

## BOILING MATERIALS FOR SERVICE AT 750-1100°

Symbol	C	Mn	Si	S	P	Ni	Cr	Mo	W	V
B <sub>4</sub>	0.35-0.45	0.50-0.80	.....	0.50 max.	0.04 max.	1.50-2.00	0.50-0.80	0.30-0.40	.....	Ni-Cr-Mo
B <sub>6</sub>	0.35 max.	0.30-0.50	0.50 max.	0.03 max.	0.03 max.	.....	4-6	0.45-0.65*	0.75-1.25*	5 Cr
B <sub>8</sub>	0.12 max.	0.60 max.	0.50 max.	0.50*	0.14*	.....	11.50-13	.....	.....	13 Cr (free mach.)
B <sub>7</sub>	0.35-0.45	0.60-0.90	0.15-0.25	0.05 max.	0.04 max.	.....	0.80-1.10	0.15-0.25	.....	Cr-Mo
B <sub>7a</sub>	0.35-0.45	0.60-0.90	0.15-0.25	0.05 max.	0.04 max.	.....	0.80-1.10	0.45-0.65	.....	Cr-High Mo
B <sub>11</sub>	0.40-0.60	0.20-0.40	0.15-0.30	0.04 max.	0.04 max.	.....	1.00-1.50	.....	1.70-2.30	W-Cr-V
B <sub>12</sub>	0.35-0.45	0.60-0.90	.....	0.05 max.	0.04 max.	1.00-1.50	0.45-0.75	.....	.....	Ni-Cr
B <sub>13</sub>	0.30-0.40	0.60-0.90	0.15-0.30	0.04 max.	0.04 max.	.....	0.45-0.75	0.40-0.65	0.85-1.35	W-Mo-Cr
B <sub>14</sub>	0.35-0.50	0.40-0.70	0.15-0.30	0.05 max.	0.04 max.	.....	0.80-1.10	0.30-0.40	.....	Cr-Mo-V
B <sub>15</sub>	0.40-0.50	0.40-0.70	0.50-0.80	0.03 max.	0.03 max.	.....	1.00-1.50	0.40-0.60	0.20-0.30	Sr-Cr-Mo
B <sub>8</sub>	0.07 max.	0.20-0.70	0.75 max.	0.50*	0.14*	7-10	17-20	.....	.....	18 : 8 (free mach.)

\* Either Mo or W may be used.

\* When P is over 0.045, S shall not exceed 0.05. When S is over 0.05, P shall not exceed 0.045.

The ferritic steels are to be heat treated, including drawing back at least 100° above operating temperature. 18 : 8 to be reheated and rapidly cooled. (2000° water quenched, for example.)

## PROPERTIES—BOLTS 2½ IN. AND UNDER. OIL QUENCHED AND DRAWN

Steel	Draw Temp.	Tensile	Yield	Elong. %	R.A. %
B <sub>4</sub>	1000	160,000	135,000	14	45
	1200	125,000	105,000	16	50
	1000	120,000	95,000	15	45
B <sub>5</sub>	1200	100,000	85,000	17	55
	1000	150,000	125,000	13	45
B <sub>6</sub>	1200	105,000	85,000	17	55
	1000	135,000	115,000	15	50
B <sub>7</sub> and B <sub>7a</sub>	1200	105,000	90,000	17	55
	1000	210,000	190,000	11	35
	1200	185,000	165,000	13	40
B <sub>11</sub>	1000	125,000	105,000	16	50
	1200	105,000	85,000	18	55
B <sub>12</sub>	1000	155,000	130,000	14	45
	1200	120,000	100,000	16	55
B <sub>13</sub>	1000	145,000	120,000	14	45
	1200	125,000	105,000	16	50
B <sub>14</sub>	1000	160,000	135,000	14	45
	1200	125,000	105,000	16	50
B <sub>15</sub>	1200	125,000	105,000	16	50
18-8	....	75,000	30,000	35	50

## A 158-37T

## ALLOY STEEL PIPE FOR SERVICE AT 750-1100°

Ferritic steels to be full annealed or normalized and drawn 200° above service temperature, except P5c, which is heated at 1350° for stabilization. Properties: tensile, 60,000; yield, normalized 30,000, full annealed 25,000, 30% Elong., 2-in. longitudinal, 25% transverse.

Steel	C	Mn	Si	S	P	Cr	Mo	W	Ti	Cb
P <sub>3a</sub>	0.15 max.	0.40-0.60	0.45-0.75	0.05 max.	0.04 max.	1.50-2.00	0.60-0.80			
P <sub>4b</sub>	0.15 max.	0.30-0.60	0.50 max.	0.03 max.	0.03 max.	1.75-2.25	0.45-0.65			
P <sub>5a</sub>	0.15 max. or 0.20 <sup>a</sup>	0.50 max.	0.50 max.	0.03 max.	0.03 max.	4-6	0.45-0.65	0.75-1.25 <sup>z</sup>		
P <sub>5b</sub>	0.15 max.	0.30 max.	1.00-2.00	0.03 max.	0.03 max.	4-6	0.45-0.65			
P <sub>5c</sub>	0.10 max. or 0.15 <sup>a</sup>	0.50 max.	0.50 max.	0.03 max.	0.03 max.	4-6	0.45-0.65			
P <sub>6</sub>	0.12 max.	0.50 max.	0.50 max.	0.03 max.	0.03 max.	12-15	2.50-3.50			
P <sub>11</sub>	0.15 max.	0.30-0.60	0.50-1.00	0.05 max.	0.04 max.	1-1.50	0.45-0.65			
P <sub>15</sub>	0.15 max.	0.30 max.	1.15-1.65	0.045	0.04	.....	0.45-0.65			

<sup>a</sup> As specified.<sup>z</sup> Either Mo or W may be used.

## Austenitic Steels.

						Ni	Cr	Ti	Cb
	0.07 max.	0.20-0.70	0.75 max.	0.03 max.	0.03 max.	8-10	17-20		
	0.08 max.	0.70 max.	0.50 max.	0.03 max.	0.03 max.	9-12	17-19		
P <sub>cc</sub>	0.25 max.	0.50-0.70	1.00-3.05	0.03 max.	0.03 max.	8-10	17-19	min.*	min.*

\* Either Ti or Cb may be used.

It will be noted that bolting stock carries much higher C than the tube steels.

**S.A.E. Steels.** White and Clark<sup>13</sup> have given creep data at 1000° for some S.A.E. and other steels for bolting stock heat treated to 285 Brinell. The Mo content of the S.A.E. steels is lower than that in analogous A.S.T.M. compositions for bolting stock.

Steel	C	Mn	Cr	Ni	V	Mo	Treatment	Grain Size	CREEP AT 1000°	
									1% 100,000 Hr.	10,000 Hr.
6140	0.40	0.75	0.21	1.01	.....	0.17	1600° oil 1200° draw		1400	3,900
3140	0.36	0.60	0.26	0.56	1.33	.....	1500° oil 1000°		190	1,025
4615	0.14	0.53	0.28	.....	1.85	0.25	1725° air 1200°		4600	7,700
4140	0.40	0.60	0.25	0.92	.....	0.18	1525° oil 1125°		3700	7,400
Ni-Cr-Mo	0.39	0.68	0.27	0.72	1.74	0.32	1525° oil 1125°	7-8	.....	780
Cr-Si-Mo	0.45	0.49	0.62	1.20	.....	0.52	1725° air 1150°		7800*	23,000*

\* These figures represent the hump in Fig. 29 and are questioned by some.

Comparison of the last three emphasizes the unpredictability of creep behavior and the care that must be taken in arguing from analogies.

Another composition for which high creep properties are claimed after normalizing and drawing at 1300° is 0.45 C, 0.75 Si, 1.25 Cr, 0.95 W, 0.30% V. Another is 0.35 C, 0.20 Si, 0.80 Mn, 0.50 Cr, 1.20% W. Some of the W may be replaced by Mo. Reported creep properties on this steel vary considerably and are probably influenced by grain size. A large amount of the Cr-W steel is in satisfactory service.

Stewart<sup>70</sup> comments favorably on steel of 2 Cr, 0.80 Si, 0.50% Mo for high-temperature bolts and studs up to 1000°.

English practice is reported <sup>71</sup> to utilize oil-quenched and tempered steels of:

C	Si	Mn	Ni	Cr	Mo
0.40	0.25	0.60	3.25	1.25	....
0.28	0.25	0.60	2.40	0.50	0.50
0.30	0.25	0.60	2.00	0.50	0.60
0.32	0.25	0.60	3.00	1.00	0.55
0.43	0.20	0.60	0.20	1.30	0.85

**Steels for Low-temperature Service.** As the temperature drops, the hardness, the tensile and yield strengths, and the endurance of steel increase.<sup>72, 73, 74</sup> Static ductility is ordinarily somewhat lowered, though not to the level which ordinarily accompanies such a change in hardness when brought about by other means.<sup>72</sup> On the basis of static tests one would conclude that steel has a better combination of properties at low than at room temperature. However, if low-temperature, notched-bar impact tests are made, a very marked falling off in impact resistance may be noted. The low-temperature brittleness of steels, a phenomenon related to stress concentration, is brought out clearly only by tests on notched specimens. With ordinary as-rolled structural steel the sharp drop in impact comes only a little below room temperature, as Fig. 23, Chap. 1, Vol. I, shows. The superiority of the Al-treated, fine-grained "Izett" steel is clearly shown in that figure.

**Evaluation of Low-temperature Impact Properties.** The usual difficulty besets the engineer in fixing an acceptance limit for impact values for a particular service, since one cannot say that anything above a certain value is good and anything below it bad. The tendency seems to be to consider 10 ft.-lb. Charpy at the temperature of service to be a minimum below which unsafe brittleness would be present. The required minimum will vary with the degree in which shock enters the service conditions. Ordinary rail steel<sup>75</sup> may show only 2 ft.-lb. Izod on a notched bar and as low as 10 on an unnotched bar at  $-20^{\circ}$ , yet failures are infrequent.

Some confusion also results because of the various dimensions of and notches in, the impact specimens used, as has been discussed in Chap. 1, Vol. I. A comparison of the Izod bar, with V-notch and the Charpy bar, with the keyhole notch, in low-temperature tests of a fine-grained C steel, is shown in Fig. 31. The V-notch shows more marked loss of toughness than the keyhole notch in tests from  $30^{\circ}$  down to around  $-10^{\circ}$ , but the latter shows an abrupt falling off around

$-30^{\circ}$ , the two branches of the curve not being connected, while the V-notch gives one smooth curve.

The selectivity of the V-notch is therefore not superior when real brittleness is met. The Charpy type of test is more convenient to make at low temperatures than the Izod, since the Charpy bar is merely laid on supports and not clamped, thus saving time and allowing the breaking of precooled bars without appreciable increase in temperature. Since the Charpy data on record have usually been obtained on bars with the regular Charpy keyhole notch, and much information is thus available for correlation, it does not seem necessary to adopt V-notches

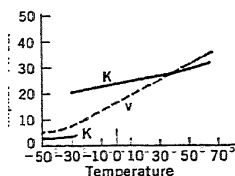


FIG. 31.—Impact results on fine-grained, annealed, 0.35% C steel, tested with two notches. V, Izod, V notch. K, Charpy, keyhole notch. (Sergeson and Poole.<sup>89</sup>)

for low-temperature impact testing, though some investigators have thought it wiser to do so.

**Effect of Deoxidation and Grain Size.** Early literature on low-temperature impact was very confusing. A given alloy steel studied by one observer would show poor results and the conclusion would be drawn that the alloying element was harmful. Another observer would find good results and conclude that the element was beneficial. The work of Herty and McBride<sup>76</sup> cleared the situation. They compared C steels deoxidized in various ways and hence with varying grain-coarsening propensities, ranging from 0.14 to 0.22 C, 0.43 to 0.71% Mn. One of these contained 0.25% Ni. The room temperature impact on as-rolled stock varied from 95 to 140 "Charpy-Izod" (Charpy bar with Izod V-notch) for the lower C killed or semi-killed steels down to 40 for a killed steel of 0.22% C and 20 for a 0.14% C rimmed steel. At  $0^{\circ}$  the rimmed steel gave but 3, the others 5 to 10, with the exception of the Ni-containing steel, which gave 32. After normalizing, the Al-treated steels gave 120 at  $0^{\circ}$  and 90 to 120 at  $-35^{\circ}$ , and the others were somewhat improved. The as-rolled condition was examined for some of the steels by the hole-tensile test at  $32^{\circ}$  (a hole is punched in a plate tensile specimen and brittleness is shown by low elongation). The one (low-Ni) steel that was tough at  $0^{\circ}$  by the impact test was brittle in this test, whereas another C steel that was brittle in the impact test was ductile in  $\frac{3}{4}$ -in. plate but brittle in 1-in. plate. Herty and McBride concluded that as-rolled C steels, even

though of relatively high coarsening temperature, are likely to show very poor low-temperature toughness. The ferrite network (produced at the boundaries of the original austenite grains as hypoeutectoid steel cools through the critical range) is evidently a source of weakness in impact tests at low temperatures. If the network is very thin, as is the case when the austenitic grain size is small, the impact resistance is much better than when it is thick. Fine grain is therefore a prerequisite in steel for low-temperature service.

#### Effect of Vanadium or Aluminum in Medium Manganese Steel.

The virtue of controlling grain size by additions of Al or V is clearly brought out by low-temperature tests. Considering first only moderately low temperatures such as may be met in the northern United States in winter, grain-size control offers a means for material improvement. This can be seen from some data on medium-Mn steels.

Hopkins <sup>77</sup> reports the following data on plates:

C %	Mn %	Si %	V %	Tensile	Yield	Elong. % 2 in.	R. A. %	CHARPY	
								10°	-25°
0.09	0.43	0.02	.....	46,500	24,500	44.5	75.5	52	4
0.16	1.16	0.16	0.19	68,000	50,000	30.5	63	35	37
0.24	1.20	0.03	.....	.....	.....	.....	.....	39-47	3-24

Heimke and Schulte <sup>78</sup> studied normalized and annealed C, medium-Mn and Cr-Mn-Si steels. From these data we may take the following for normalized specimens:

C %	Mn %	Si %	Cr %	CHARPY	
				70°	-20°
0.25	0.49	0.01	....	27	4
0.25	1.24	0.06	....	31	18
0.26	1.32	0.44	....	45	35*
0.20	1.05	0.67	0.42	40	4.5
0.23	1.21	0.68	0.45	39	12
0.09	1.31	0.91	0.45	51	33*

\* Fine-grained steels.

The embrittling effect of high C and high Mn together will, of course, exert itself at some composition, in spite of fine grain, that is, composition must also be taken into account.

**Effect of Phosphorus.** Sergeson's <sup>79</sup> curves include a steel of 0.22 C, 0.07 Si, 1.04 Mn, and 0.08% P which, normalized, gave 25 Charpy at



+75° and 5 at -25°. The high P was thought to be responsible to a major extent for the low impact value. (See also Chap. 2.)

With due consideration to the embrittling effect of the C, Mn, and P and other alloy content, so that such elements are held to proper limits, most of the alloy steels can be made serviceable at moderately low temperature by grain-size control.

Herzig and Parke<sup>50</sup> show Figures 32 and 33 for a variety of S.A.E. steels in the normalized and quenched and drawn conditions.

In the normalized condition, the superiority of the lower carbon steels is evident. The plain high carbon normalized steel, although almost the softest of the series, is the weakest in impact, the highly alloyed steels, especially 4340 which air-hardened considerably on normalizing, are the next weakest. The good behavior of the Ni steel 2330 is evident.

In the quenched and tempered condition, drawn back to 300 Brinell, the low-carbon steels are again superior, and an important point emerges in the great difference between the fine-grained and the coarse-grained steels. It is unfortunate that the grain size established by the quenching temperature actually used was not given instead of the McQuaid Elrn grain size, since the two may be different but the effect of the different propensities toward coarsening is plain.

The table shows that the steels of this investigation did not harden fully on quenching. The authors\* state that the steels were quenched in large sections, and the impact specimens taken from the centers.

An extended report on low-temperature properties has been published by the A.S.T.M.† since the earlier printing of this volume and should be consulted for much additional information.

**Effect of Nickel.** For services at very low temperatures, as in solvent dewaxing plants in the oil industry, Ni steels are widely used, since Ni has a specific effect in raising the low-temperature impact resistance. Figure 34 from Aldridge and Shepherd<sup>51</sup> shows this effect. Even at such temperatures some fine-grained steels without Ni show good results, as Fig. 35 illustrates. Kinzel<sup>52</sup> states that a steel of 0.10 C, 0.24 Mn, 0.24 Si, 0.83 Cr, and 0.53% Cu, giving S5 Izod at room temperature, showed 92 at -60° and 41 at -110°.

It is obvious that no true comparison can be made of the effect of

\* Private communication.

† Special Research Committee Report, Project 13—Impact Resistance and Tensile Properties of Metals at Subatmospheric Temperatures. American Society for Testing Materials, August 1941, 112 pages.

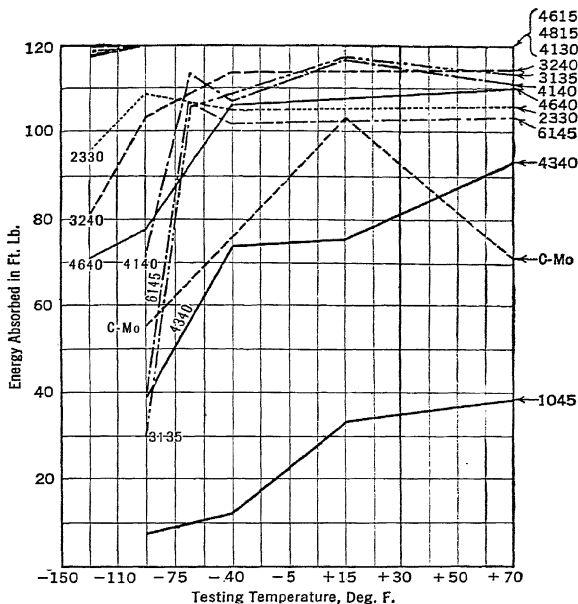


Fig. 32.—Impact properties (Charpy bars with V-notch) of 225 Brinell S.A.E. steels at temperatures shown. (Herzig and Parke.<sup>50</sup>)

SAE No.	Normalized	BHN	Grain Size by Mc-Quaid-Ehn Test
4130	1650	235	8
4140	1650	255	8
4615	1630	192	7-8
4640	1700	235	8
4815	1700	202	8
4340	1630	440	8
C-Mo*	1700	137	3-8 Duplex
2330	1675	192	7
3135	1700	223	4-5
3240	1675	293	6-7
6145	1650	286	7-8
1045	1650	174	1-8 Duplex

\* Analysis = C 0.15%; Mn 1.00%; Si 0.22%; Mo 0.47%.

Heat Treatment and Hardness				Draw Temp.		Draw Temp.	
SAE No.	Temp., Deg.	Quenching	Rockwell "C" After Quench*	Deg.	BHN	Deg.	BHN
4130	1575	Water	48.5	1075	269	1260	207
4140	1575	Oil	42.5	1075	286	1240	228
4615	1525	Water	42.5	940	294	1100	235
4640	1525	Oil	44.5	1040	302	1240	217
4815	1475	Oil	37.5	800	302	1100	228
4340	1525	Oil	45.0	1100	302	1260	228
C-Mo*	1625	Water	36.0	650	309	1100	228
2330	1450	Oil	40.0	875	286	1050	217
3135	1500	Oil	45.0	875	318	1200	217
3240	1475	Oil	44.5	1000	286	1240	228
6145	1525	Oil	44.5	1125	286	1250	228
1045	1525	Water	40.0	475	---	940	228

\* The specimens were taken from the center of large bars, which accounts for the low hardness as quenched.

Data for Figs. 32 and 33.

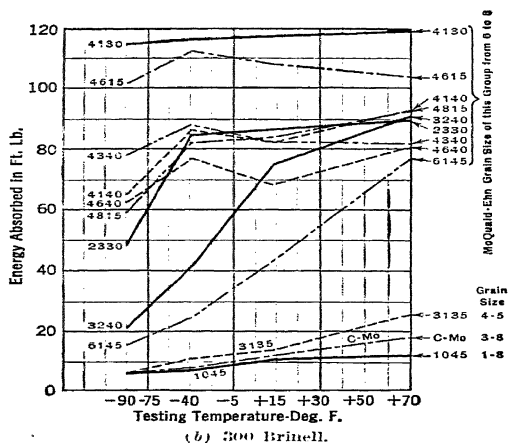
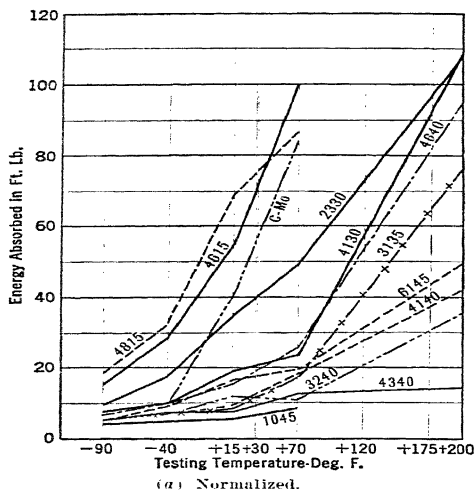


FIG. 33.—Impact properties at temperatures shown of some S.A.E. steels, quenched and tempered to two ranges of hardness. (Herzig and Parke.<sup>50</sup>)

alloying elements on low-temperature brittleness without first making sure that the steels compared have the same grain size. Clear evidence for superiority of steels high in Ni at extremely low temperatures, such as that of liquid air, is, however, on record, and has been summarized by Russell.<sup>72</sup> Since Ni is a ferrite strengthener one might suspect that its action would be to render a ferrite network less harmful, but of all the ferrite strengtheners its effect seems most potent. The effect in austenite is equally powerful. The low-temperature properties of the austenitic ferro-nickels, and of 18 Cr, 8 Ni are superior, as Russell and Welcker<sup>73</sup> and Malcolm<sup>83</sup> have shown. Since the other useful prop-

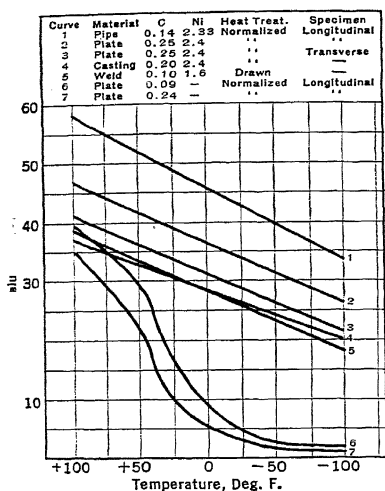


FIG. 34.—Effect of temperature on the impact resistance of Ni and C steels. (Aldridge and Shepherd.<sup>81</sup>)

For data on other Ni cast steels at sub-zero temperatures see reference 95.

erties of 18 : 8 are not brought into play at sub-zero temperatures, the cost factor usually leads to the selection of ferritic steels for low-temperature service.

**Summary.** Bolting stock, for use at such moderately elevated temperatures that oxidation is not a problem, is made from deep-hardening alloy steels of the usual medium constructional types, quenched and tempered well above the temperature of operation.

For high-temperature service the heat treatment is designed to produce a steel stable against the tempering effect of long high-temperature service. This is usually accomplished by normalizing, following by drawing well above the temperature of service. A lamellar pearlitic

structure has higher load-carrying ability than an annealed or spheroidized structure; hence the first type, although theoretically less stable, is preferred.

The C content is not as critical for high-temperature strength as it is in steels for lower temperature service. It is doubtful if raising the C above some 0.30–0.40% adds strength, and the increase in strength on raising it from 0.10 to 0.20% is not very great. However, the steels of, say, 0.07–0.20% C are superior in strength to the carbonless alloys. As a rule, low C is desired in order to avoid embrittlement resulting from the solution of carbides at elevated temperature and

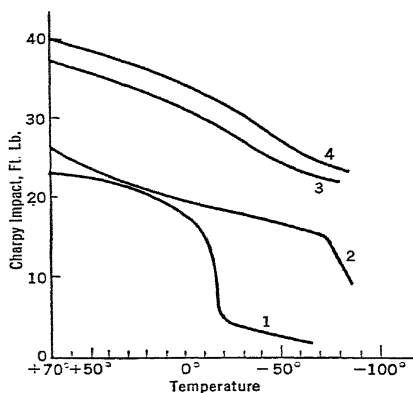


FIG. 35.—Low-temperature impact of V steels. (Vanadium Steels and Irons.<sup>55</sup>)

No.	Composition, %				Normalized from	Drawn at
	C	Mn	Ni			
1	0.26	0.70	...		1625°	
	0.35	0.90		0.20	1800°, 1560°	
	0.29	1.00	1.54	0.10	1775°, 1475°	
	0.31	1.65		0.10	1650°, 1500°	750°

their precipitation at a lower range of temperatures. Binding the C with an excess of Ti or other strong carbide former, may be resorted to in order to avoid embrittlement.

Molybdenum, a specific against temper brittleness in quenched and tempered steels, is useful in combating embrittlement in high-temperature service. The resistance to tempering that is characteristic of Mo heat-treated steels finds its counterpart in the marked strengthening of steels at high temperature. Molybdenum in amounts of 0.50%, or more, is added to practically all the pearlitic steels for service from 750 to 1300°.

Tungsten has a similar effect, but at least double as much W as Mo is required. Since W is the more expensive, the choice generally falls on Mo. For oxidation and corrosion resistance, Cr is added, in amounts ranging from about 1.50 to about 9% in pearlitic steels, and to 15 to 25% in austenitic alloys. Silicon or Al, or both, are added in varying, but much smaller, quantities, both to pearlitic and austenitic steels.

The austenitic steels have superior load-carrying ability at the higher range of temperatures, often being used for service above 1200°, and almost universally for service above 1300°. Molybdenum or W are added, up to around 5%, to the austenitic type, to confer increased load-carrying ability.

The austenitic steels, as well as the lower Cr steels, are, as a class, susceptible to carbide precipitation and resulting embrittlement, which must be combated by suitable precautions.

The effect of austenitic grain size in the pearlitic steels upon the creep resistance is very marked, and the better performance of coarse-grained than of fine-grained steels is evident, at least in C steels, as low as 750°. Coarsening the grain to produce improved creep resistance involves danger of loss of impact resistance, but a suitable compromise can probably be worked out in most cases, though this has so far not been given much attention. Steels made fine grained by addition of V seem likely to be better in creep resistance than those of equally fine grain produced by Al additions.

In steels for low-temperature service, fine grain is sought, since the impact resistance of coarse-grained steels is usually very low. The creep resistance and the impact resistance, respectively, gained in coarse- and fine-grained steels may, of course, not be directly due to grain size, but may relate to factors in the making of the steel, particularly the deoxidation practice, which are likewise reflected in the grain-coarsening characteristics of the steels.

For service at extraordinarily low temperatures, Ni is an effective addition, and is usually considered as much of a specific for this purpose as Mo is for improving high-temperature strength.

In high-temperature service, stability is paramount. Surface stability, i.e., resistance against oxidation and corrosion, and structural stability, i.e., retention of load-carrying ability for a long period, and avoidance of carbide separation and resulting embrittlement are sought, so the steels are built up with suitable alloying elements and the heat treatments chosen with stability in view.

Much high-temperature service requires tubular products, so ability to be formed into tubes is a factor in selection of compositions. In

cast products, the ease of making sound castings of uniform properties is a factor in the selection, which accounts for the use of higher C and higher Ni in heat-resisting castings than are common in wrought products.

The field of steels for high- and low-temperature service are special ones in which expensive alloying elements are used. The steels are tailored to fit the service, and standardization is not at all complete. There is still much room for improvement in these steels, and much need for added information on the properties of those already known and used.

### BIBLIOGRAPHY

1. DIXON, E. S.: Needs of the Oil Industry for Metals at High Temperatures, *A.S.T.M.-A.S.M.E. "Symposium on Effect of Temperature on the Properties of Metals,"* 1931, pp. 66-99.
2. TAPSELL, H. J.: "Creep of Metals." London, Oxford University Press, 1931, 285 pages.
3. CROSS, H. C.: High Temperature Tensile, Creep, and Fatigue of Cast and Wrought High- and Low-carbon 18 Cr 8 Ni Steel from Split Heats, *Trans. Am. Soc. Mech. Eng.*, v. 56, 1934, pp. 533-554.
4. BAILEY, R. W.: Creep and Engineering Design, *International Assoc. Testing Materials*, London, April, 1937, pp. 15-17.
5. ABORN, R. H., and R. F. MILLER: Ductility of Creep-resistant Steel at Elevated Temperatures, *Metals & Alloys*, v. 9, 1938, p. 104.
6. BAILEY, R. W.: Steel at Elevated Temperatures, *J., West Scotland Iron Steel Inst.*, v. 45, 1937-38, pp. 11-12.
7. WHITE, A. E., C. L. CLARK, and R. L. WILSON: The Rupture Strength of Steels at Elevated Temperatures, *Trans. Am. Soc. Metals*, v. 26, 1938, pp. 52-80.
8. American Society for Testing Materials: Cooperative Short-Time High Temperature Tension Tests of Carbon Steel K6, *Proc. Am. Soc. Test. Mat.*, v. 33, part 1, 1933, pp. 213-217.
9. KANTER, J. J., and L. W. SPRING: Some Long-Time Tension Tests of Steels at Elevated Temperatures, *Proc. Am. Soc. Test. Mat.*, v. 30, part 1, 1930, pp. 110-132.
10. American Society for Testing Materials: The Determination and Significance of "Proportional Limit" and "Breaking Strength" in Short-Time High-Temperature Tests, *Proc. Am. Soc. Test. Mat.*, v. 33, part 1, 1933, pp. 218-224.
11. CLARK, C. L., and A. E. WHITE: The Properties of Metals at Elevated Temperatures, *Univ. of Michigan, Engineering Research Bulletin* 27, 1936, 100 pages.
12. HATFIELD, W. H.: Permanence of Dimensions under Stress at Elevated Temperatures, *J. Iron Steel Inst.*, v. 122, 1930, pp. 215-247.
13. WHITE, A. E., and C. L. CLARK: Comparison of Single-Step Long-Time Creep Results with Hatfield's Time-yield Stress, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 481-504.

14. PILLING, N. B., and R. WORTHINGTON: The Effect of Temperature on Some Properties of Iron-Chromium-Nickel Alloys, *A.S.T.M.-A.S.M.E.* "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 495-556.
15. McVETTY, P. G.: The Interpretation of Creep Tests, *Proc. Am. Soc. Test. Mat.*, v. 34, part 2, 1934, pp. 105-122.
16. FOLEY, F. B.: Discussion, *Metals & Alloys*, v. 6, 1935, pp. 50-51.
17. WHITE, A. E., and C. L. CLARK: Influence of Grain-Size on the High Temperature Characteristics of Ferrous and Nonferrous Alloys, *Trans. Am. Soc. Metals*, v. 22, 1934, pp. 1069-1098.
18. KANTER, J. J.: Discussion, *A.S.T.M.-A.S.M.E.* "Symposium on Effect of Temperature on the Properties of Metals," 1931, p. 237.
19. NORTON, F. H.: "The Creep of Steel at High Temperature," New York, McGraw-Hill Book Co., 1929, 90 pages.
20. BAILEY, R. E., J. H. S. DIXON, N. P. INGLIS, and J. L. PEARSON: The Trend of Progress in Great Britain on the Engineering Use of Metals at Elevated Temperatures, *A.S.T.M.-A.S.M.E.* "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 218-244.
21. ROBINSON, E. L.: Effect of Temperature Variation on the Creep Strength of Steels, *Trans. Am. Soc. Mech. Eng.*, v. 60, April, 1938, pp. 253-259 (R.P. 60-5).
22. MONTGOMERY, H., and J. W. BOLTON: Furnace for Elevated Temperature Tests, *Metals & Alloys*, v. 5, 1934, p. 127-128, 130.  
GILLET, H. W., and H. C. CROSS: Obtaining Reliable Values for Creep of Metals at High Temperatures, *Ibid.*, v. 4, 1933, pp. 91-98, 104.  
Tentative Method of Test for Long-Time (Creep) High Temperature Tension Tests of Metallic Materials, *Am. Soc. Test. Mat. Tentative Standards*, 1937. E 22-35 T. pp. 1544-1555.
- McVETTY, P. G.: Temperature Measurement and Control in Creep Testing. *Am. Soc. Test. Mat.*, 1938. Advance Copy.
23. American Society for Testing Materials and American Society of Mechanical Engineers. "Compilation of Available High-Temperature Creep Characteristics of Metals and Alloys," Phila. and New York, Either Society. 1938, 848 pages.
24. WILSON, R. L.: High Temperature Strength of Steels, *Metal Progress*, v. 33, May, 1938, pp. 499-505.
25. CROSS, H. C., and E. R. JOHNSON: Creep Properties of 5 Per Cent Chromium, 0.50 Per Cent Molybdenum Steel Still Tubes, *Proc. Am. Soc. Test. Mat.*, v. 34, part 2, 1934, pp. 80-104.
26. CROSS, H. C., and F. B. DAHLE: High Temperature Properties of Cast and Wrought Carbon Steels from Large Valves for High Temperature Service, *Trans. Am. Soc. Mech. Eng.*, v. 58, 1936, pp. 103-114.
27. CROSS, H. C., and J. G. LOWTHER: "Study of Effects of Manufacturing Variables on the Creep Resistance of Steels." Appendix to Report of Joint Research Committee. To be published in *Proc. Am. Soc. Test. Mat.*, v. 38, part 1, 1938.
28. CLARK, C. L., and A. E. WHITE: Discussion, *A.S.T.M.-A.S.M.E.* "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 382-389.
29. JARES, V.: Discussion, *International Assoc. Test. Mat.*, London Congress, April, 1937, p. 182.
30. WHITE, A. E., C. L. CLARK, and R. L. WILSON: Qualifications of Steels for



- High-temperature Service in Petroleum Refinery Equipment, *Oil & Gas J.*, v. 33, August 2, 1934, pp. 16, 18-19, 51.
- WHITE, A. E., C. L. CLARK, and R. L. WILSON: Factors Influencing Creep of Alloy Steels Used in Refinery Equipment at High Temperature, *Ibid.*, v. 33, August 9, 1934, pp. 16, 33, 34, 36.
31. KINZEL, A. P.: Discussion, *A.S.T.M.-A.S.M.E.* "Symposium on Effect of Temperature on the Properties of Metals, 1931, pp. 389-390.
  32. JENKINS, C. H. M., H. J. TAPSELL, G. A. MELLOR, and A. E. JOHNSON: Some Aspects of the Behaviour of Carbon and Molybdenum Steels at High Temperatures, *Trans. Chem. Eng. Congress of the World Power Conference*, London, 1936, v. 1, pp. 122-162.
  33. WEAVER, S. H.: Actual Grain Size Related to Creep Strength of Steel at Elevated Temperatures, *Am. Soc. Test. Mat.*, 1938, advance copy.
  34. Allowable Working Stresses for Ferrous Materials, *Mech. Eng.* v. 60, 1938, pp. 170-171.
  35. NEWELL, H. D.: Properties of the 9% Chromium Steel (an intermediate alloy), *Metal Progress*, v. 29, February, 1936, pp. 51-55.
  36. MILLER, R. G., R. F. CAMPBELL, R. H. ABORN, and E. C. WRIGHT: Influence of Heat Treatment on Creep of Carbon-Molybdenum and Chromium-molybdenum-silicon Steel, *Trans. Am. Soc. Metals*, v. 26, 1938, pp. 81-105.
  37. CROSS, H. C., and D. E. KRAUSE: Phosphorus as an Alloying Element in Steels for Use at Elevated Temperatures, *Metals & Alloys*, v. 8, 1937.
  38. FLEISCHMANN, M.: Selection of Steels for High Temperatures, *Steel*, v. 102, January 17, 1928, pp. 34-39.
  39. FLEMING, W. R.: Steel Alloy, U.S. Patent 2,051,991, August 25, 1936.
  40. MOCHEL, N. L.: Low Carbon Stainless Steels, Forgings and Rolled Bars (in Thum, E. E.: "The Book of Stainless Steels," 2nd ed., pp. 251-266).
  41. NEWELL, H. D.: Effect of Silicon on Chromium-molybdenum Steels for High-temperature Service, with a note on the Effect of Copper, *Metals Technology*, v. 4, September, 1937, 15 p. Also *Am. Inst. Min. Met. Eng.*, Tech. Publ. 835 (1937).
  42. KOHLE UND EISENFORSCHUNG G.m.b.H.: British patent application 12,515 (1937).
  43. DAUVERGNE, J.: Les Tubes de chauffage dans les raffineries de pétrole (Heating Tubes in Petroleum Refineries) *Rev. de Mét., Mém.*, v. 34, 1937, pp. 676-689.
  44. FLEISCHMANN, M., and S. D. WILLIAMS: Aluminum Alloy Steels for Sour Crude Refining, *National Petroleum News*, v. 29, July 28, 1937, pp. R51-55.
  45. GOTTIGNIES, M. L.: Discussion, *International Assoc. Test. Mat.*, London Congress, April 19-24, 1937, pp. 184-185.
  46. BAILEY, R. W.: Discussion, *International Assoc. Test. Mat.*, London Congress, April 19-24, 1937, pp. 182-183, 185-186.
  47. COMSTOCK, G. F., and C. L. CLARK: Effect of Titanium on Some Properties of 17.5% Chromium Steel, *Metals & Alloys*, v. 8, 1937, pp. 42-46. See also BANNON, R. E.: Effect of Titanium on the Hardness and Microstructure of Heat Treated 18 Per Cent Chromium Steel Ingots, *Trans. Am. Soc. Metals*, v. 25, 1937, pp. 737-754.
  48. HATFIELD, W. H.: Heat-resisting Alloys, *J. Inst. Fuel*, v. 11, 1938, pp. 245-304.
  49. TUCKER, W. A. and S. E. SINCLAIR: Creep and Structural Stability in Nickel-Chromium-Iron Alloys at 1600° F., *Bureau Standards, J. Research*, v. 10, 1933, pp.

50. ROSENHAIN, W., and C. H. M. JENKINS: Some Alloys for Use at High Temperatures. 1. Nickel-chromium and Complex Iron-Nickel-Chromium Alloys, *J. Iron Steel Inst.*, v. 121, 1930, pp. 225-236.
51. KANTER, J. J.: See, Thum, E. E.: "The Book of Stainless Steels," 2nd Ed., 1935, p. 437.
52. NEWELL, H. D.: Performance in High Temperature Service (in Thum, E. E.: "Book of Stainless Steels," 2nd ed., 1935, pp. 405-418).
53. "Die Heraeus-vacuumschmelze." Hanau-am-Main, G. M. Alberti, 1933, 434 pages.
54. KRIVOBOK, V. N.: A Digest of Stainless Steels, *Yearbook Am. Iron Steel Inst.*, v. 27, 1937, pp. 129-167.
55. BROWN, R. S.: Discussion, *A.S.T.M.-A.S.M.E.* "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 93-98.
56. HATFIELD, W. H.: Discussion, *International Assoc. Test. Mat.*, London Congress, April 19-24, 1937, pp. 188-190.
57. OERTEL, W., and A. SCHEPERS: Hitzebeständiger Stahl durch Legierung mit Chrom, Nickel, Aluminium, und Silizium (Heat Resistant Alloys by Alloying with Chromium, Nickel, Aluminum and Silicon) *Stahl u. Eisen*, v. 52, 1932, pp. 511-513.
58. WOLFSON, S. I., and A. M. BORZDIKA: Short Time Creep Characteristics of Stainless Steels, *Metals & Alloys*, v. 8, 1937, pp. 294-296.
59. BECKET, F. M.: Chromium-Manganese Steels, *Yearbook Am. Iron Steel Inst.*, v. 20, 1930, pp. 173-194.
60. MONYPENNY, J. H. G.: Chromium-manganese Steels as Heat-resisting Steels. *Metallurgia*, v. 17, 1938, pp. 93-96.
61. SCHMIDT, M., and H. LEGAT: Hitze beständige Chrom-Mangan-Stähle (Heat Resistant Chromium-Manganese Steels), *Archiv f. Eisenhüttenw.*, v. 10, 1936, pp. 297-306.
62. VANICK, J. S., W. W. DE SVESHNIKOFF, and J. G. THOMPSON: Deterioration of Steels in the Synthesis of Ammonia, *U. S. Bureau Standards*, Tech. Paper 361, v. 22, 1927, pp. 199-233.  
VANICK, J. S.: Deterioration of Steel and Wrought Iron Tubes in Hot Gaseous Ammonia, *Trans. Am. Soc. Steel Treat.*, v. 4, 1923, pp. 62-78.  
VANICK, J. S.: Deterioration of Some Metals in Hot Reducing Ammonia Gases, *Proc. Am. Soc. Test. Mat.*, v. 24, part 2, 1924, pp. 348-372.
63. KOSTING, P. R.: Deterioration of Chromium-tungston Steels in Ammonia Gases, *Metals & Alloys*, v. 5, 1934, pp. 54-56.
64. MAXWELL, H. L.: The Effect of Gases on Ferrous Materials at High Temperatures and High Pressures. *Pennsylvania State College, Mineral Industries Experiment Station*, Bull. 18, 1935, pp. 128-136.
65. RANQUE, G.: La méthode d'autostabilisation thermique à vitesse d'allongement imposée et l'établissement de diagrammes caractéristiques de fluage, *Rev. de Mét.*, Mém., v. 34, 1937, pp. 349-360.
66. MOCHEL, N. L.: The Seizing of Metals at High Temperatures, *Proc. Am. Soc. Test. Mat.*, v. 28, part 2, 1928, pp. 269-277.
67. BOLTON, J. W.: Seizure of Metals at Elevated Temperatures, *Mech. Eng.*, v. 58, 1936, pp. 165-168.
68. ROBINSON, E. L.: A Relaxation Test on 0.35 C Steel (K 20), *Trans. Am. Soc. Mech. Eng.*, v. 59, 1937, pp. 451-452.

69. MOCHEL, N. L.: Relaxation Tests on 0.35 C Steel (K 20) at 850° F., *Ibid.*, v. 59, 1937, pp. 433-455.
70. STEWART, W. C.: High Temperature Properties for Some Alloys of Particular Interest to the Navy, *J. Am. Soc. Naval Eng.*, v. 50, 1938, pp. 107-132.
71. HATFIELD, W. H.: Creep as Occurring in Different Steels under Service Conditions, *International Assoc. Test. Mat.*, London Congress, 1937, pp. 21-23. See also *Engineering*, v. 144, 1937, p. 78.
72. RUSSELL, H. W.: Effect of Low Temperatures on Properties of Metals, *A.S.M.E.-A.S.T.M.*, "Symposium on Effect of Temperature on the Properties of Metals," 1931, pp. 658-812.
73. RUSSELL, H. W., and W. A. WELCKER, JR.: Damage and Overstress in the Fatigue of Ferrous Metals, *Proc. Am. Soc. Test. Mat.*, v. 36, part 2, 1936, pp. 118-138.
- WISHART, H. B., and S. W. LYON: Effect of Overload on the Fatigue Properties of Several Steels at Various Low Temperatures, *Trans. Am. Soc. Metals*, v. 25, 1937, pp. 690-701.
- WISHART, H. B., and W. D. BOONE: Properties of Rail Steel at Arctic Temperatures, *Metal Progress*, v. 29, January, 1936, pp. 61-62.
- HERTY, C. H., JR., and D. L. MCBRIDE: Effect of Deoxidation on the Impact Strength of Carbon Steels at Low Temperatures, *Cooperative Bull.* 67, Min. Met. Advisory Boards to Carnegie Inst. Tech. and Bur. Mines, Pittsburgh, 1934, 43 pages.
- HOPKINS, R. K.: Impact Resistance of Some Steels and Welds at Sub-zero Temperatures, *J. Am. Welding Soc.*, v. 13, October, 1934, pp. 16-28.
- HEIMKE, H. W., and W. C. SCHULTE: Low Temperature Impact Tests of Medium Manganese Steel Plate, *Metals & Alloys*, v. 5, 1934, pp. 31-36.
- SERGEON, R.: Behavior of Some Irons and Steels Under Impact at Low Temperatures, *Trans. Am. Soc. Steel Treat.*, v. 19, 1932, pp. 368-384.
- HERZIG, A. J., and R. M. PARKE: Laboratory Investigation of Low Temperature Impact Properties of Some S.A.E. Steels, *Metals & Alloys*, v. 9, 1938, pp. 90-93.
81. ALDRIDGE, B. G., and G. C. SHEPHERD, JR.: Nickel Steels at Low Temperatures, *Ibid.*, v. 7, 1936, pp. 147-152, 185-192.
82. KINZEL, A. B.: Machine Element, U. S. Pat. 2,018,267, October 22, 1935.
83. CLARK, P. H., and E. L. ROBINSON: An Automatic Creep Test Furnace Guide, *Metals & Alloys*, v. 6, 1935, pp. 46-51.
84. CROSS, H. C., and J. G. LOWTHER: Report on Long-Time Creep Tests of 18 Per Cent Chromium, 8 Per Cent Nickel Steel and 0.35 Per Cent Nickel Steel and 0.35 Per Cent Carbon Steel, *Proc. Am. Soc. Test. Mat.*, v. 37, part 1, 1937, pp. 177-186.
85. McVETTY, P. G.: New Equipment for Creep Tests at Elevated Temperatures, *Ibid.*, v. 37, part 2, 1937, pp. 235-257.
86. WRIGHT, E. C.: Five Per Cent Chromium Steels; As Rolled Material, Especially Tubing (in Thum, E. E.: "The Book of Stainless Steels," 2nd Ed., 1935, pp. 234-244).
87. CROSS, H. C., and J. G. LOWTHER: Long-Time Creep Test of 0.35 Per Cent Carbon Steel (K 20). Appendix to Report of Joint Research Committee. To be published in *Proc. Am. Soc. Test. Mat.*, v. 38, part 1, 1938.
88. Vanadium Corporation of America: "Vanadium Steels and Irons." New York, The Author, 1937, 189 pages.

- SERGESON, R., and S. W. POOLE: Cooperative Study of Low-temperature Impact Testing. *Proc. Am. Soc. Test. Mat.*, v. 36, part 1, pp. 132-142.
90. Timken Steel and Tube Co. "Digest of Steels for High Temperature Service," November 1, 1934. The Company, Canton, O.
91. Babcock and Wilcox Tube Co. "Digest of Properties, Carbon and Alloy Steels for Cracking-Still Tubes." Tech Bull. 6a, 1935. The Company, Beaver Falls, Pa. See also *Metals & Alloys*, v. 7, March, 1936, p. 73.
92. MALCOLM, V. T.: The Effect of Low Temperature on Properties of Metals. *Refrigerating Eng.*, v. 35, 1938, pp. 189-193.
93. THIELMANN, R. H., and E. R. PARKER: The Fracture of Steel at Elevated Temperatures. Paper presented at the October, 1938, meeting of the Am. Inst. Min. Met. Eng.
94. WHITE, A. E., C. L. CLARK, and C. H. MCCOLLAM: Influence of Chromium, Silicon and Aluminum on the Oxidation Resistance of Intermediate Alloy Steels. *Am. Soc. Metals*, Preprint 9, October, 1938, 23 pages.
- Castings for Sub-zero Service. *Nickel Steel Topics*, v. 7, October, 1938, p. 8.
- NAUMANN, F. K.: Manufacture of Articles from Steel Alloys. U. S. Pat. 2,132,877, October 11, 1938.

## THE AUSTENITIC STEELS

**Formation of Austenite.** A sufficient addition of a depth-hardening element, Cr for example, may make austenite so sluggish that, at readily attainable cooling rates, considerable amounts of austenite are retained in quenching, even though the martensite transformation cannot readily be entirely suppressed, so that 100% austenite does not result from the addition of most of the alloying elements.

The elements Mn and Ni can, however, be added in such large amounts, that, even in the absence of C, the gamma-alpha transformation is depressed far below room temperature, or even entirely avoided at any temperature. Thus steels of 14% Mn or 30% Ni, without C, are austenitic. One per cent C lowers the necessary amount of Mn to 6% and of Ni to 12%.

In ternary steels, containing some other elements that make austenite sluggish, these required amounts are reduced, e.g., 0.60 C, 14 Ni, 15 Cr, 2% W; 0.70 C, 15 Cr, 5% Mn, or 0.10 C, 18 Cr, 8% Ni are austenitic. The effect of C on the gamma loop in Fe-Cr alloys has been noted in Fig. 3, Chap. 7. The composition may be such that the austenite can be transformed only under certain conditions, i.e., so that the gamma-alpha transformation is not reached under ordinary conditions, but nevertheless is attainable at very low temperature, or can be induced by heavy pressure. When the latter condition exists, the alloys are resistant to certain types of wear.

**Properties of Austenite.** The austenitic steels of major importance are those containing large amounts of Mn, for wear resistance, and those with large amounts of Cr (plus some other element) for heat and corrosion resistance. The austenitic steels are characterized by great toughness, of a somewhat different type from that of ferritic steels, since the austenites show greater elongation and less necking-down in the tensile test.

Austenite has high tensile but low elastic strength. However, the material work hardens rapidly, so that the elastic strength of cold-worked austenite may be brought up close to the ultimate. The material has great ability for deformation but requires more energy for deformation than do ferritic materials.

In a completely austenitic steel there is no possibility of grain refinement due to passing through an allotropic change, so the only way to refine the grain is by working. Slow-cooled castings have a strong tendency to produce very large grains. The total surface at the grain boundaries is less with large than with small grains, and the boundaries are important in connection with precipitation-hardening effects. These factors affect the behavior of austenitic castings.

**Heat Treatment in Respect to Carbide Solution and Precipitation.** A stable austenitic steel is not heat treatable in the ordinary sense of the word. Nevertheless, the C-containing alloys are often given a quench or air cool from a high temperature, in order to minimize the occurrence of a very irritating precipitation-hardening propensity of such alloys. In an austenite containing a carbide-forming element such as Mn or Cr, the homogeneous solid solution of metals and C has a tendency to allow the formation of chemical compounds, i.e., carbides of Mn or Cr. This formation is at a maximum at a relatively low range of temperatures, varying with composition, but ordinarily falling within the limits 900–1400°.

**Effect of Carbide Separation.** Homogeneous austenite, retained as such by quenching, shows characteristic ductility and toughness. If it is too slowly cooled through the dangerous range (900–1400°), or reheated into it and held there, carbide formation occurs, and these carbides have a strong tendency to collect at the grain boundaries. Since the carbide-forming metal (e.g., Mn or Cr) enters into these carbides, it is commonly postulated that the austenite near them is depleted both of C and of the metal, and thus, at the depleted areas, behaves like austenite of lower alloy content. In a Cr-containing austenite, the grain boundaries are thus thought to have lower Cr and to be therefore less corrosion-resistant than the body of the grain, unless ample time and temperature are given for diffusion. Another hypothesis is that the separation of the carbides at the grain boundaries produces a state of local stress that is responsible for grain boundary corrosion, and contributory to embrittlement.

The deposition of brittle carbides at the grain boundaries, especially when the grains are large and the total boundary surface small, tends to envelop the tough austenite with a honeycomb of brittle carbide and thus develop brittleness of the whole structure. If there is sufficient C and a small enough surface boundary so that the network or honeycomb is continuous and thick, the embrittlement may be very serious. Reducing the percentage of C, making the grain size small, or avoiding a sojourn in the dangerous temperature range, are obvious ways of evading the difficulty. It is also sometimes possible, when the struc-

ture has been broken up by working, to control the precipitation of carbides so that it occurs within the grains instead of being localized at the grain boundaries.

**Stabilization.** Still another procedure is to add an element that is a much stronger carbide former, in sufficient amount to combine with all the C and leave none free for combination with Mn or Cr. If these new carbides are not themselves soluble and precipitable, the embrittlement can be avoided. Titanium and Cb, and to a lesser degree Mo, W, and V, form much more stable carbides than do Mn and Cr. These stable carbides have a tendency to be peppered throughout the grains instead of being concentrated at the grain boundaries, and thus, like most inclusions within ductile grains, to have but little effect on mechanical properties.

In the absence of such stabilizers, the C of this high-Cr austenite tends to behave like any precipitation-hardening element in steel. At a high enough temperature it is in solution. On cooling rapidly, it will stay in solution. When the cooled solution is reheated, formation and separation of carbide take place, with exactly the same type of time-temperature relations for separation and agglomeration as were shown in Chap. 8, Vol. I, for precipitation hardening in general. If the temperature is raised sufficiently, the carbides again go into solution and the damage that has been done by their separation is repaired.

The separation of the compound does not necessarily require a reheating. There is an urge for separation to occur at a definite temperature range and unless the austenite, by quenching or the attainment of an equivalent rapid cooling rate by air cooling of thin sections, is forced to pass quickly through this range, separation of the compound will occur on cooling, just as it would on reheating the quenched material into this range.

**Welding.** In the welding of C-bearing austenite, the metal adjacent to the weld is likely to cool slowly enough to allow carbide separation in the zone that has a sufficient sojourn in the dangerous range of temperatures, with consequent embrittlement. The precipitation of carbides also impoverishes the matrix of this zone in the metal that enters into the carbide and for this reason is thought to favor intergranular corrosion near the weld.

**Austenitic Nickel Steels.** Nickel is not a carbide former; so the austenite in Fe-C-Ni alloys is not beset by carbide-precipitation effects. The austenitic range of Ni alloys contains many interesting alloys such as Invar, Permalloy, etc., of special physical properties. The possibilities and peculiarities of these would require much space for presentation. Discussion of them and of what heat treatments they may

be subjected to is outside the scope of this volume. The reader is referred to "Nickel Alloy Steels"<sup>1</sup> and "The Alloys of Iron and Nickel."<sup>2</sup>

We shall confine our attention to the austenitic steels of chief commercial importance, i.e., the Mn steels and the 18 Cr-8 Ni type.

**High Manganese Steels.\*** In general, the requirements for producing an austenitic commercial Mn steel necessitate a Mn content from about 6 or 8% to 20%, in combination with the proper amount of C. Below the lower limits given, the steel, even with the most suitable treatment, may be characterized by the presence of brittle martensite. The upper limits are determined by the cost of the Mn additions, and further, by the influence of the C content (when the Mn rises to around 20%), which will make the steel brittle when cold. Most commercial Mn steels have about 11 or 12% Mn and about 1.0 to 1.2% C. The more recent steels have a tendency toward a lower percentage of Mn than that originally thought necessary. Similarly, the field for the use of high-Mn steels has also been considerably broadened.

The peculiar merit of these steels lies in their resistance to abrasion rather than in any particularly high static or fatigue strength. In this regard, Mn steels appear to resist the abrasive wear characteristic of heavy impacts of hard substances better than that caused by the sliding attrition of hardened parts, or that of an abrasive wheel.

Aside from the dynamic strength, the selection of a Mn steel for any specific service depends upon the correlation of wearing qualities and static properties. In general, for maximum wear resistance the most ductile steel which will give a yield strength sufficiently high to avoid distortion in service will be best. These properties, in turn, depend upon the proper combination of C and Mn. Thus, a steel with 9 to 11% Mn and the proper amount of C will have a higher yield strength than a steel with more than 11% Mn. Again, a steel with 11% Mn and 1.10% C will have a higher yield strength than one with 15% Mn and 0.8% C. With high Mn and low C, steels quenched in water from 1830° will give a low yield strength and a flow of metal which may prove excessive for many duties. A great deal depends also upon a suitable heat treatment of the steel.

**Cast vs. Wrought Manganese Steels.** As might be expected from the influence of the rate of cooling upon the structure of high-alloy steels, the physical properties of these high-Mn steels are greatly modified by the method of casting, the size of the casting, and, in wrought material, by the mechanical work. The first two factors, in particular, have a great influence upon the toughness of the metal. The average

\*The austenitic Mn Steels are often termed "Hadfield Mn steels," since they were developed and commercialized by Hadfield.



tests of commercial Mn steels, with about 11 or 12% Mn and a little over 1.0% C, give approximately the following:

Condition of the Metal	Tensile	Yield	Elong. %
Cast .....	82,000	45,000	30
Rolled .....	135,000-140,000	60,000-70,000	30-40
Forged .....	142,000	55,000	38

The yield strength of some sections, as rolled, may even go as high as 75,000; the proper heating and working of the metal plays a very important part in the results to be obtained on physical test.

A common specification for Mn steel rails is as follows:

CHEMICAL	PER CENT
C.....	0.95- 1.15
Mn.....	10.00-13.00
Si.....	0.20- 0.40
P.....	Under 0.10
S.....	" 0.06

PHYSICAL	
Tensile.....	100,000
Yield.....	55,000
Elong. %.....	20

**Alloying of Manganese Steels.** Small amounts of Cr are sometimes added to increase the yield strength, so that in rolled sections the yield may often be as high as 85,000. It is stated that the resistance to shock does not appear to be lowered by the addition of Cr up to 1.00% but with Cr above 0.50%, elongation is rapidly decreased, and with Cr above 1.00% elongation falls below 20%. It will be brought out below that a Cu addition remedies this. A Ni-Mn steel has also come into some commercial use.

**Heat Treatment of Manganese Steels.** The heat treatment of high-Mn steels presents a most important phase in connection with the successful application of these steels. Incorrect treatment is responsible for many of the failures which have been registered against Mn steels, and usually has been caused either by a mistaken idea of the particular structure best suited to the specific work in hand, or by a lack of sufficient knowledge of the mechanics of the austenite transformation. The use of the microscope, and a judicious consideration and application of

the results obtained, are probably the best means of solving a given problem in connection with the adjustment of heat treatment.

The heat treatment for these steels involves homogenization of austenite. (As martensite is not naturally present in the principal Mn steels now used in commercial practice, and is not generally wanted, its consideration may be omitted.)

**Structure.** Commercial high-Mn steel, as cast, is fundamentally austenitic; the crystals are often excessively large, and in many instances form a weak, columnar structure with a carbide network be-



FIG. 1a.—Commercial manganese steel annealed at 1750°  $\times 100$ . (Bullens.)

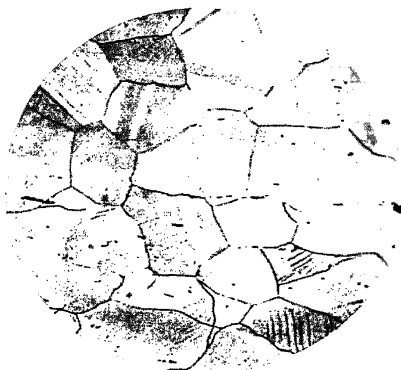


FIG. 1b.—Rolled manganese steel water quenched from 1925°  $\times 100$ . (Battelle.)  
C, 1.36; Si, 0.36; Mn 14.20%

tween them. It is evident that such a steel, for many purposes, will be entirely unsatisfactory.

On the other hand, many high-Mn steels, as forged, are characterized by an exceedingly fine, almost chalky structure, and yet may be very brittle. If a bar of Mn steel be heated to 1800°, and one-half be allowed to cool slowly and the other half quenched in water, both ends will have a comparatively fine grain size; yet the slow-cooled end will have but 2 to 4% elong. as against 50 to 60% elong. in the quenched end. This effect of annealing is due to the formation, on very slow cooling, of the maximum amount of carbide, an extremely hard and brittle Mn-containing cementite, rejected by the austenite by a precipitation effect. This substance forms as a weak membrane around the austenite grains, as spines and needles, or in some other characteristic manner. This is shown in the micrograph of Fig. 1a, taken from a rolled commercial Mn steel, annealed at 1750°.

**Annealing Not a Satisfactory Treatment.** The effect of annealing upon the physical properties is shown by the following table of results, obtained from tests upon wrought annealed Mn steels:

HIGH-MANGANESE STEELS, ANNEALED

C %	Mn %	Tensile	Yield	Elong. %	R.A. %
0.95	10.07	94,600	67,500	1	1.5
1.00	11.21	99,950	77,660	1	0.75
1.07	13.38	103,670	62,350	3	3.5

Annealing these high-C Mn steels is obviously illogical, and it is the carbide which is the main source of the difficulty. On heating to a high temperature, the carbide membranes produced by slow cooling, or the carbide segregations, are gradually taken into solution by the austenite; and, by rapid cooling from that same temperature, the carbide is entirely or almost entirely prevented from reprecipitating, and especially from taking on the weakening structure (that of a membrane or of segregations) previously mentioned. Since the carbide originally formed in casting is very sluggish in being absorbed by the austenite (this is a more or less marked characteristic of all hypereutectoid steels), and as the homogenization of the steel as a whole also takes place slowly, a high temperature is necessary. Furthermore, the temperature must be high, and the cooling must be effected very rapidly—as by water quenching—to retain the carbide in solution. Such a treatment will give the most ductile steel. The temperature required is generally not less than 1830°. Thus, a water quenching, from 1830°, of the wrought annealed steels previously considered will show a tensile strength of about 135,000 to 145,000, with 50 to 60% elong. Figure 1*b* shows the structure of a properly quenched Mn steel.

On the other hand, by varying the factors of temperature, duration of heating, and rate of cooling, it is possible to obtain physical properties covering a wide range. The static strength and ductility are largely governed by the amount of the original free carbide which is taken into solution and there retained by water quenching. Thus, the properties of the steel, looking at the results of heat treatment from this point of view, may be varied from those characteristic of the steel as cast, rolled, or forged, to those indicative of a full "water toughening."

High-Mn steel has no critical or transformation points. Thus, whereas in the ordinary steels the heat treatment is more or less guided

by such temperatures, in high-Mn steels the only criterion of proper temperatures is the relation of the carbide to the physical properties; the absorption, with or without the precipitation, of such carbide is the underlying basis for adjustment of heat treatment.

**Nickel in Hadfield Steel.** One interesting recent development is the addition of 3 to 5% Ni to welding rod.<sup>3</sup> Welds in austenitic Mn steel, made in the usual way, with rod similar in composition to the casting, are weak at the edge of the weld, owing to overheating of the parent metal and to carbide precipitation in the affected zone. Although the use of a different welding rod would not have been expected to alter the behavior of the zone next to it, in arc welding, the Ni-bearing rod, preferably also of lower C than the casting, is found to blend with the parent metal sufficiently so that the weld does not materially injure the adjacent metal. The weld holds, and carbide separation near the weld is maintained at so low a level as to be far less injurious.

Payne<sup>4</sup> remarks that Ni helps the weld, but a line of weakness still exists in the parent metal. Spraragen and Claussen<sup>5</sup> review the literature of welding cast Mn steel.

Hall<sup>6</sup> also finds the Ni-Mn austenitic steels useful for rolling into rod and drawing to wire, such as is used in heavy screens that must resist impact abrasion. The usual water quenching is not necessary to make the rolled Ni-containing steel ductile; air cooling will serve. The following properties are reported:

	Rough Cast	Cast, Ground	Rolled, 1-in. rod
Tensile.....	97,000	105,500	138,500
Yield.....	27,000	27,000	41,00
Elong. %.....	30	57	83
R.A. %.....	23	48	50

Increased toughness at elevated temperature is claimed<sup>7</sup> for a steel of 0.65 C, 12.50 Mn, 4% Ni, air cooled from 1920°.

**Chromium in Hadfield Manganese Steel.** Chromium, and both Cr and Cu, may be added to Hadfield steel. The comments in the paragraphs below on the effect of such additions are taken from information kindly supplied by the Electro Metallurgical Co.\*

For several years a number of manufacturers have been adding from about 1.00 to 1.50% Cr to Hadfield Mn steel used in the crushing, grinding, or conveying of light-weight, small-size material such as gravel and sand, where the wear is primarily abrasive, with lack of

\* Private communication, February 14, 1938.

sufficient impact readily to develop the skin hardness that is essential for the obtaining of maximum life in the plain Mn steel. Some care must be taken in the production of Cr-bearing Hadfield Mn steel. If the C content is above about 1.30%, the steel is likely to be somewhat brittle, particularly if the Mn-C ratio is not kept at the optimum, generally considered to be about 11 to 1. An addition of less than 0.75% Cr seems to produce no worthwhile improvement in wear resistance, and it would appear that 1.50% is about the maximum that it is desirable to add.

The Cr appears to improve wear resistance of Hadfield Mn steel in two ways:

1. It slightly increases the hardness before any work hardening is done on the steel. A plain Mn steel will have a Brinell hardness of approximately 200, one containing 1.50% Cr, 220, and one containing 2.00% Cr will have, according to Clark and Coutts,<sup>8</sup> a Brinell hardness of 286.

2. Less cold work or deformation is necessary to develop the required skin hardness.

Not many data are available on steels of this type. The following results on commercial heats are of interest:

Heat	C %	Mn %	Cr %	Tensile	Yield	Elong. %	R.A. %	B. H. N.
1.....	1.28	12.90	1.54	119,000	74,500	30	22	228
2.....	1.38	13.05	1.44	115,500	76,500	18.5	15	240

Tests with a higher Cr content gave the following:

Heat	C %	Mn %	Si %	Cr %
1037 1039 1045	1.00 *	12.00 *	0.50 *	5.00 *

Bar	Quench	Tensile	Yield	Elong. %	R.A. %
1037A	1740°	133,500	78,500	25	23
1037B	"	114,500	64,500	11.5	15
1039A	"	131,500	78,000	11	9.5
1045A	"	111,500	72,500	16	13.5
1045B	1790°	125,000	71,500	19	17.5

\* Nominal.

**Copper and Chromium in Manganese Steel.** This steel, although capable of being forged, was more difficult to forge than a steel of similar composition but containing Cu as well as Cr.

The effect of Cu on a steel of this type is shown by the following data:

Heat	C %	Mn %	Si %	Cr %	Cu %
1038	1.00	12.00	0.50	5.00	0.50
1046	"	"	"	"	"
1032	"	"	"	"	1.00

Bar	Quench	Tensile	Yield	Elong. %	R.A. %
1038A	1740°	140,500	77,000	12	12.5
1038B	"	118,000	71,600	11.5	10
1046A	"	137,000	75,000	18.5	15.5
1046B	1790°	134,500	76,000	22	20
1032A	1690°	125,000	67,500	35	25
1032B	1740°	132,000	70,500	33.5	28
1032C	"	140,500	75,500	38	34

These tests indicated that the addition of 1% Cu to a Hadfield Mn steel containing 5% Cr was definitely helpful in increasing the ductility. However, the steels containing 5% Cr were in general somewhat inferior to those containing only 3% Cr, as the following shows:

Heat	C %	Mn %	Si %	Cr %	Cu %
947	1.34	13.48	0.90	1.16	1.25
958	1.20*	12.00*	0.50*	2.00*	1.50*
959	1.20*	12.00*	0.50*	3.00*	1.50*

Bar	Quench	Draw	Time	Tensile	Yield	Elong. %	R.A. %
947A	2000°	.....	.....	161,000	70,000	63	29
947C	"	750°	6 hr.	153,000	70,000	59.5	39
958A	"	.....	.....	140,000	65,000	54	39
958B	"	750°	6 hr.	136,500	65,000	49	46
959A	"	.....	.....	153,500	71,000	52	40
959B	"	750°	6 hr.	152,000	73,500	48	36.5
959C	"	800°	3 hr.	154,000	68,500	50	36.5

\* Nominal.

Hence, the addition of 3% Cr and 1.50% Cu is helpful in Hadfield Mn steel which is to be used under conditions causing flow in the standard grade, such as in rails and frogs. The C should be kept between 0.80 and 1.10%, excepting for applications where ductility is not of great importance, when it may be raised to 1.30-1.40%.

**Effects of Various Alloying Elements.** Krainer<sup>9</sup> examined forged Mn steels with a variety of C and Mn contents, and with several alloying elements. Selected data from his tables are given below. The steels were water quenched from 1800 to 1830°.

		Mn	Ni	Cu	Other	Tensile	Yield	Elong.	R.A.	Brin
0.61	Low	13.61				140,000	51,000	52		
1.20		13.61				168,000	60,000	81	54	
0.64		16.78				156,500	54,000	73	58	209
1.27		16.72				165,000	68,000	49	216	
0.68		19.21				138,000	52,500	58	200	
1.40		19.34				166,500	67,000	58	41	253
1.10		13.60				164,000	58,500	81	47	211
1.04		14.16				158,000	62,500	86	57	209
1.00		15.31				158,000	71,000	50	39	231
1.10		17.85				165,000	71,000	65	44	234
1.04	Low	14.75				156,000	61,000	86	54	210
1.00		16.54				146,500	59,500	84	58	218
1.03		19.43				149,000	58,000	84	50	195
1.06		15.40				160,500	62,500	71	58	225
1.02		14.40		1.60		162,000	72,500	42	33	268
1.03		19.96		4.20		153,500	84,000	30	24	271
1.04		13.79			1.06	155,000	61,000	89	54	201
1.08		16.92			2.07	151,000	58,500	82	54	208
1.03		19.81			2.04	145,000	61,000	71	52	198
0.96		19.80				159,000	61,000	65	44	217
0.90		18.84			2.00 Mo	159,000	65,500	72	45	215
1.03		17.32			0.27 V	169,000	72,500	52	36	235
0.93		17.45			0.52 V	172,000	86,500	40	33	245
1.10		16.97			1.00 Ta	159,000	68,000	44	36	265
0.98		12.36			2.00 Ti	143,500	67,000	35	31	248
1.10		15.84			1.00 Ti	159,000	64,000	57	42	218
0.96		18.83			2.00 Ti	141,000	72,500	37	31	229
		17.94			0.25 Zr	151,000	58,500	60	42	212
		18.13			1.23 Zr	151,000	67,000	40	33	236

The effect of V in raising the yield strength is very marked. Krainer studied the work-hardening ability of the alloys, which he considered as a criterion of wear resistance, and found Cr to be effective in increasing this.

**Wear and Fatigue Resistance.** The wear-resistant property of austenitic Mn steel rests upon its great propensity toward work hardening. Though the material is soft to a Brinell test, i.e., it flows easily, and is soft to the first strokes of a file, a Rockwell test made in a Brinell impression where the steel has been work hardened, or the subsequent strokes of the file, find it very hard. Under pressure and dis-

tortion, a hard skin forms, which is often considered to be martensite, but may perhaps be chiefly severely cold-worked austenite, and this skin is really responsible for the wear resistance. Light abrasive pressure, insufficient to form the skin, results in rapid wear.

Basing his conclusions on experience with pins in chain in severe service, Cauley<sup>10</sup> states that austenitic Mn steel is not well suited for service under repeated stress, owing to cracking of the work-hardened surface and development of fatigue failures from the cracks as nuclei.

**Austenitic Nickel-Chromium Steels.** The well-known 18 Cr, 8% Ni type of austenitic steel is of value for its corrosion resistance, its resistance to scaling at high temperature, its strength at high temperature, and its formability. Volumes have been written on this alloy and

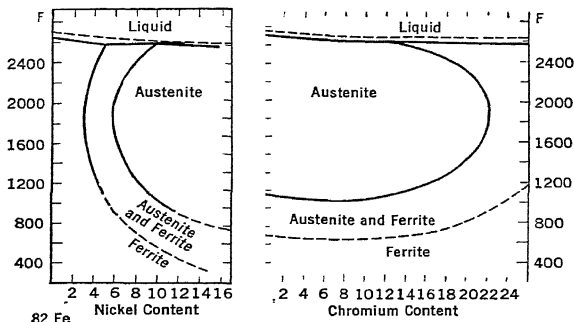


Fig. 2a.—Approximate constitutional diagrams for 18% Cr alloys with variable Ni, and for 8% Ni alloys with variable Cr.

its modifications.<sup>2, 6, 11, 12, 13</sup> From the point of view of the general heat treater, the discussion can be condensed here, since the principles involved are similar to those involved in handling austenitic Mn steel. The specialist who must consider very fine gradations must look elsewhere for the necessarily complicated and detailed discussion.

The 18 : 8 alloy is not an entirely stable austenite. Cold working of low-C 18 Cr, 8 Ni type alloys, in which Ni is at the lower limit, produces alpha iron, as is shown by the fact that the magnetic permeability, negligible in the annealed or quenched material, is vastly increased by severe cold work. This is analogous to the hardening of austenitic Mn steel by cold work. Decrease in either Ni or C favors the development of permeability, i.e., formation of ferrite at lower amounts of cold deformation. The tendencies of Ni, Cr, and C toward ferrite formation are shown in Fig. 2. The dashed portions of the lines indicate that ferrite is not ordinarily found in that region, but that it



can exist. Increase in Ni makes appearance of ferrite more difficult. This is one reason for the use of higher Ni in many of the cast alloys.

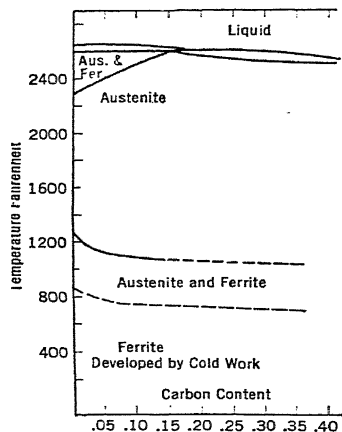


FIG. 2b.—This chart shows the approximate transformation temperatures for the austenite-ferrite equilibrium in 18 Cr 8% Ni steels. (Aborn and Bain.<sup>15</sup>)

**Effect of Nickel.** The effect of increasing Ni upon the constitution of 18% Cr steels is shown in Fig. 3 from Krivobok and Grossmann.<sup>14</sup> The effect of increasing Ni in extending the austenite field and of de-

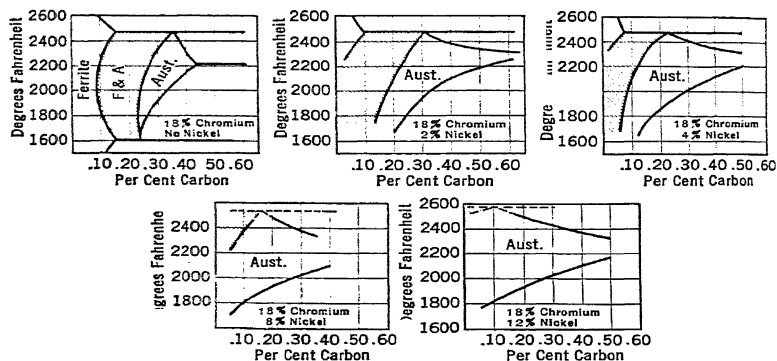


FIG. 3.—Diagrams illustrating the effect of Ni in modifying the constitutional diagram for 18% Cr steels. (Krivobok and Grossmann.<sup>14</sup>)

creasing the shaded area in which both ferrite and austenite exist is evident. Aborn and Bain<sup>15</sup> show this in another way (Fig. 4).

**Properties of 18:8.** As a material of construction for room temperature use, 18:8 has fair strength and good ductility. As the C increases from about 0.05 to 0.20%, in sheet that has been rapidly cooled from 2100°, the tensile rises from about 75,000 to about 95,000 with elongation and reduction both in the range of 65–75%. The proportional limit is low, ranging from about 20,000 to 30,000. In bar stock the material is very tough, giving about 95 to 120 Izod. Castings show 65,000 to 85,000 tensile, 30,000 to 45,000 yield, 25,000 to

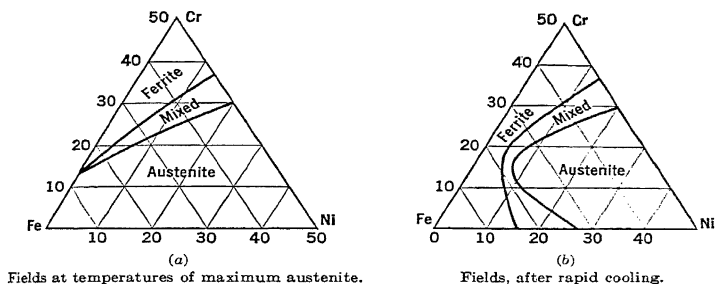


FIG. 4.—The FeNiCr fields. (Aborn and Bain.<sup>15</sup>)

30,000 proportional limit, and 40 to 65% elong. and R.A. with 90–95 Izod.

**Cold-worked 18:8.** Cold-worked material may be softened, or pre-precipitated carbides taken into solution, in commercially acceptable times, by heating to 1800° or above. Half an hour at temperature at 1800° or only a minute or so at 2100° will suffice. The higher temperatures give somewhat greater softness but tend to produce grain growth. For material under 0.10% C, 1800 to 1900° is perhaps the usual range of temperature employed. The higher C alloys are more commonly heated to 2000–2100°. In cooling, the range 1500–750° should be passed through as quickly as possible, to avoid carbide precipitation. This range represents the temperatures at which, for different C contents, the carbide is no longer truly stable, but wants to come out by the precipitation-hardening mechanism. The effect of C content on the temperature required is indicated in Fig. 5.

Cold working progressively raises the strength and vastly improves the yield ratio. Tensile of 300,000 with yield of 250,000 and some remaining ductility, i.e., 3% R.A. and perhaps 1% elong., are properties attainable on very severe cold work.

Intermediate properties, such as 185,000 tensile, 150,000 yield (0.2%

set), 5% elong. in 2 in., can be had. The 18 : 8 steel cold worked to 150,000 tensile, 110,000 to 120,000 yield, is used in thin sheet, spot welded into structural shapes that have the requisite stiffness, in the production of some of the modern high-speed railway cars. Krivobok and Lincoln,<sup>16</sup> and the discussions of their paper, have shown that slight changes in Ni, Cr, and C content affect the degree of reduction in cold rolling necessary to reach this strength, and that the elongation, for this tensile strength, may vary, in the different compositions and degrees of reduction, from 8 to 60%. The formability by bending also varies, and not necessarily in direct relation to the elongation. The value for yield strength is highly dependent upon the definition, i.e.,

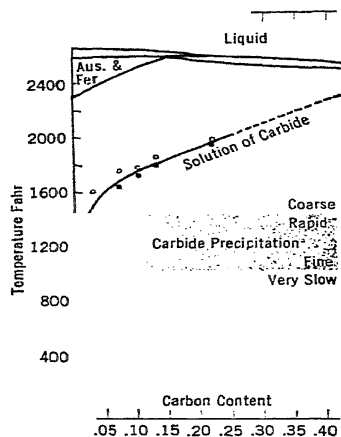


FIG. 5.—Solubility of C in the austenite of 18 Cr, 8 Ni steel. (Aborn and Bain.<sup>15</sup>)

the amount of permanent set taken as corresponding to yield strength, because of the shape of the stress-strain diagram.

**Welding of 18 : 8.** Carbide precipitation in grain boundaries has had results physically and chemically. The higher C, 18 : 8, after sojourn in the dangerous temperature range, such as occurs in welding by ordinary procedures, may become very brittle, cracking on slight bending. The precipitated carbides are high in Cr and low in C, so the matrix adjacent to the precipitate is thought to be depleted in Cr. If the Cr in the matrix at the grain boundaries is reduced below the level necessary to give corrosion resistance to the environment met, or, on the alternative hypothesis, if too much stress is introduced at the boundaries, by the precipitation there of carbide particles, the steel will corrode in intergranular fashion, in certain corrosive media.

Nevertheless, welded 18 : 8, in which a brittle zone is produced and which would corrode badly at the zone of carbide precipitation, in some corrosive media, will be satisfactory for many decorative uses in which the material is not flexed and which involve only atmospheric exposure.

**Stabilized 18 : 8.** For avoidance of carbide precipitation, addition of Ti or Cb to "fix" the C (requiring about 6 times as much Ti as C or 10 times as much Cb as C) may be resorted to. Material so alloyed is not sensitive to slow cooling through the "dangerous range"; hence, for welding, a Ti, or a Cb, addition is advantageous. Actually, the alloying of the material itself with Ti and the incorporation of Cb in the

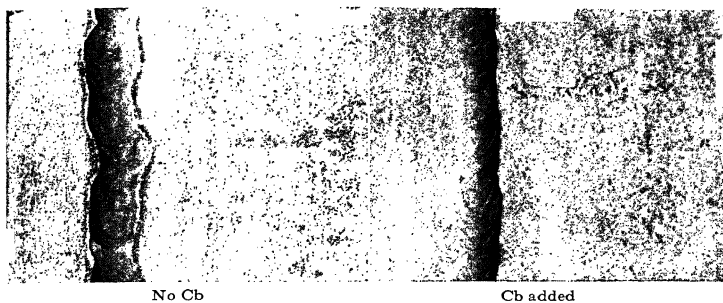


Fig. 6.—Stabilization of 18 : 8 by Columbium. (Courtesy of A. B. Kinzel, Union Carbide & Carbon Research Laboratories.)

The photograph on the left shows an arc weld made in an annealed 18 : 8, with 0.07% C, no Cb. That on the right shows the same in an annealed 18 : 8 with 0.06 C, 0.74% Cb. All plates were  $\frac{1}{8}$  in. thick. The as-welded samples were boiled for four 1-hr. periods in a solution containing 20%  $\text{HNO}_3$  and 4% HF. The metal adjacent to the weld in the steel without Cb is appreciably attacked, whereas it is practically unattacked in the steel with the Cb addition.

welding rod seem a desirable combination, because Cb is not so readily oxidized and lost in the heat of the weld as is Ti. The virtue of stabilization against carbide precipitation in welding is shown by Fig. 6.

A stabilizing treatment of 2 to 4 hr. at about  $1560^\circ$  is given to the Ti-containing 18 : 8. This seems to favor the binding of the C by the Ti so firmly that at the lower temperature of the "dangerous range" the C does not switch its allegiance to join up with Cr and to precipitate.

The embrittlement is due to carbide collection in grain boundaries; the intergranular corrodibility may be due to Cr depletion of the matrix at the boundaries. At any rate, the two phenomena do not reach their maxima simultaneously, as is indicated in Fig. 7. This figure also shows that the tendency for change increases with C. The stress accompanying precipitation tends to produce some ferrite, the steel be-

coming very slightly magnetic, just as occurs in severe cold working. Thus a third phenomenon appears. So many actions going on make the interpretation of experimental evidence in terms of theory somewhat difficult.

**High Titanium.** Thum<sup>17</sup> points out that a 0.06 C, 0.45% Ti, 1S : 8 sheet cold rolled to 160,000 tensile can be precipitation hardened (by the separation of some Ti compound) on reheating to about 925°; the properties then become 250,000 tensile, 220,000 yield, 3% elong. The material is machinable in the initial state and readily formable;

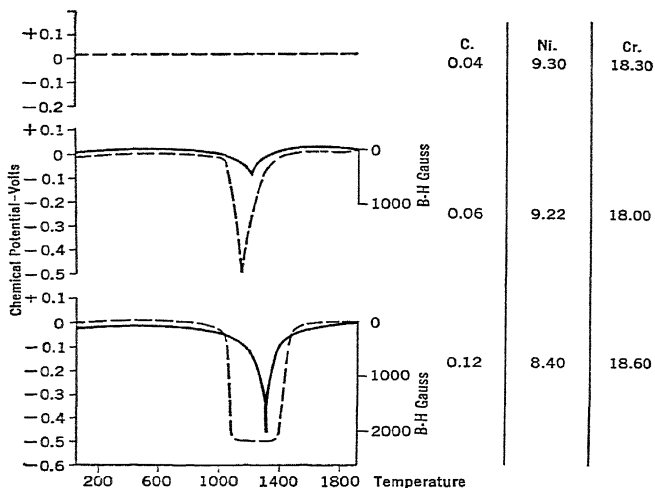


Fig. 7.—Changes in chemical potential (solid line) and magnetic behavior (dashed line) of 1S : 8 on heating 3 hr. at temperatures shown. (Strauss, Schottky and Hinüber.<sup>31</sup>)

The steel with 0.04% C showed no change in chemical potential and the change in magnetism was too small to show on the scale on which the results are plotted. Note the restoration of the original condition on heating over 1400°.

whereas, if these properties were induced by cold work without precipitation hardening, it would not be. Ability to precipitation harden is not met in this Ti-treated alloy until the Ti content is raised to some 3%, unless the material is cold worked, so that no such precipitation occurs in the welds of Ti-stabilized, unworked 18 : 8.

**Alloys in 18 : 18.** Other alloying elements used are Si, Al, W, Mo, and Cu.\* Silicon adds somewhat to strength and especially to scaling resistance. It may run up to 5%. Mo has some stabilizing effect on

\* Recent discussions<sup>32-34</sup> have been given of the effects of alloys.

the carbide, analogous to that of Ti, but its chief use is increasing corrosion resistance; around 3% adds much to the resistance to the sulphite liquors met in paper-mill practice. Smith<sup>17</sup> comments on the value of Mo in preventing the pitting type of corrosion. Up to 4% W may be added. Both W and Mo are used in 18 : 8 to confer high-temperature strength, but it is not certain that these additions have a beneficial effect comparable to that which they confer on pearlitic steels. This was discussed in the previous chapter.

Free-machining austenitic stainless steels containing S up to 0.30–0.40%, Se up to 0.20–0.30%, MnS or ZrS, are on the market. The static properties are scarcely affected by these additions, nor is the corrosion resistance vastly altered.

**Copper and Manganese in 18 : 8.** The 18 : 8 Cr-Ni composition was chosen because, of all the austenitic Cr-Ni compositions, it requires the least Ni (see Fig. 4) and is therefore the cheapest. Since Mn and Cu are both similar to Ni in many of their alloying effects, their behavior as substitutes for, or additions with, Ni has had some study. Grimshaw<sup>18</sup> describes an alloy of 0.10 C, 17.50 Cr, 8 Ni, 5 Mn, and 3% Cu. As rolled, it shows 90,000 tensile, 42% elong., 66% R.A., 170 Brinell; quenched from 1950° it gives 80,000 tensile, 52% elong., 71% R.A., 125 Brinell. The proportional limit is stated to be 6000 lb./in.<sup>2</sup> higher than in ordinary 18 : 8. The alloy can be precipitation hardened to 180 Brinell by heating between 600 and 1100°. If as-rolled sheet is drawn at 1200° prior to welding, it is claimed to be immune to intergranular corrosion after welding, although carbide precipitation is marked after the 1200° draw. The alloy does not work harden as rapidly as 18 : 8. The Mn + Cu addition is said to enhance the corrosion resistance, and to improve machinability materially. A further addition of around 2% Si may be made to increase resistance to scaling. A steel of 0.09 C, 18 Cr, 4.50 Ni, 5.50 Mn, 1% Cu, quenched from 2100°,<sup>19</sup> shows 90,000 tensile, 45,000 yield, 55% elong., 75% R.A. Günzburg and co-workers<sup>20</sup> stated that 7 Mn, 2% Cu added to 18% Cr without Ni produces an austenitic steel with about the same mechanical properties as 18 : 8, but it is doubtful that this composition is completely austenitic.

**Manganese in Steels of the 18 : 8 Type.** Replacement of Ni by Mn has been discussed by Becket<sup>21</sup> and the Fe-Cr-Mn system has recently been described. A weight-for-weight replacement of Ni by Mn does not produce the same structure in the alloys. Burgess and Forgeng<sup>22</sup> show, by Fig. 8, that stable austenite requires a minimum of approximately 16% Mn for steel of 15% Cr. The "sigma" phase which appears on slow cooling of steels with higher Cr, when the Mn is high, is brittle; hence compositions are avoided in which it would occur.

Bruhl<sup>23</sup> and Schmidt and Legat<sup>24</sup> discussed the Cr-Mn steels, and

showed the necessity of avoiding compositions giving what Burgess and Forgeng term the sigma phase. They conclude that, for avoidance of embrittlement in high-temperature service, the pure austenitic steels within this group are necessary, and that the Cr content consistent with a fully austenite structure is so low that the steels are satisfactorily resistant to scaling only below about  $1500^{\circ}$ . Only small addition of Si or Al to increase this limit can be made because Si or Al act like an increase in Cr to throw the alloy outside the pure austenite range. In discussion, Riedrich remarked that not only must the Cr not go

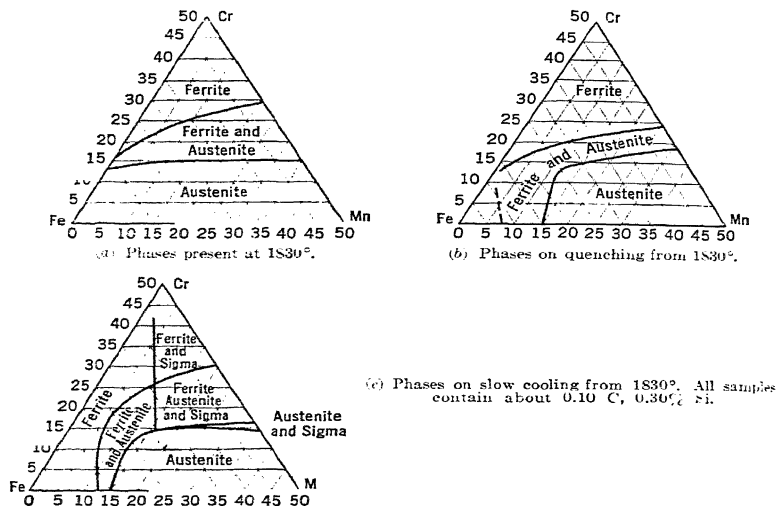


FIG. 8.—Structure of FeMnCr alloys. (Burgess and Forgeng.<sup>22</sup>)

much over 16% or the Mn below 15%, but also the C content should be low or some addition made to "fix" the C, else carbide embrittlement will appear. He suggests reducing the Cr and adding fairly large amounts of Si. He considers the stainlessness of 15 Cr, 16 Mn, 1% Mo equal to that of a straight 15% Cr steel. Rapatz, in discussion, cited 0.20 C, 17 Mn, 3 Cr, 1% Mo as fully austenitic, and thought that the Si-containing steels advocated by Riedrich would be prone to embrittlement. He rates 18 Cr, 10% Mn as equivalent to 18 Cr, 8% Ni, in corrosion resistance against several reagents, but not to resistance to sulphuric acid. At room temperature the properties of 18 Cr, 10% Mn are close to those of 18 Cr, 8% Ni except that the yield strength of the

Mn alloy is much higher, 65,000–70,000 against 35,000–50,000. The higher yield strength for the Mn alloy also holds over the range of 600–1100°, in short-time, high-temperature tests.

Hougardy and Riedrich<sup>25</sup> point out that the addition of 1.50 Ni and 0.75% Mo to a 17 Mn, 10% Cr alloy, which contains both ferrite and austenite, increases the corrosion resistance. In austenitic heat-resistant steels of 0.10–0.16 C, 16.5–19.5 Mn, 7.5–8.5 Cr, 3.6–4.3% Si they

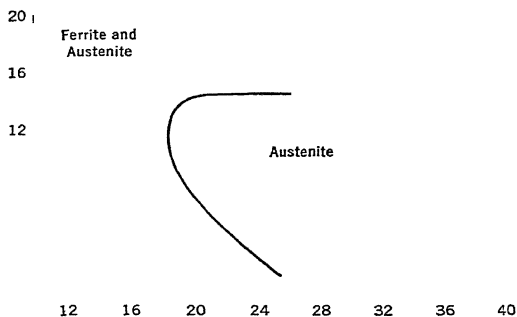


FIG. 9.—Structure of FeCrMn alloys with 0.10–0.20% C. Dashed line, limit of fully austenitic structure as forged; full line, limit for worked and reheated steels. (Legat.<sup>26</sup>)

find that, after quenching from 1920° and reheating at 1110°, the impact resistance falls from 22 mkg. cm.<sup>2</sup>, as quenched, to 2 mkg. cm.<sup>2</sup> after 50 hr. heating; whereas, if about 0.60% Ti is added, the impact, which as quenched is 19.5 mkg. cm.<sup>2</sup>, falls only to 18.5 after 100 hr. heating.

Legat<sup>26</sup> gives us Fig. 9, which agrees with Fig. 8. He made additions to steel of about 0.20 C, 14 Cr, 16% Mn with results on as-forged specimens given below:

Addition	Tensile	Yield	Elong. %	R.A. %	Impact mkg. cm. <sup>2</sup>	Brinell
None. . . . .	132,000	100,000	42	60	90	260
2.50% Ni. . .	128,000	85,000	38	62	100	235
1.00% Ni. . .	128,000	92,000	40	60	100	255
2.50% Mo. . .	137,000	100,000	40	62	100	...

The yield strengths are high for this type of material, so his as-forged samples were probably cold worked to some degree, as the hardness also indicates.



Becket<sup>21</sup> described Cr-Mn and Cr-Mn-Cu steels in 1930. Some of his data follow:

C %	Si %	Cr %	Mn %	Cu %	Treatment	Tensile	Yield	Elong. %	R. A. %	Izod	Brinell
0.07	0.04	17.88	8.26	.....	2100° air	120,000	48,000	24	22	120	163
0.05	0.27	17.88	16.53	.....	1920° water	102,000	47,000	49	69	120	174
0.22	0.41	16.93	7.62	.....	1830 "	121,000	46,000	21	31	63	267
0.25	0.29	16.13	13.64	.....	1920 "	143,000	45,000	45	41	116	183
0.10	0.52	17.20	9.78	0.52	2100 "	116,000	47,000	40	39	120	179
0.12	0.89	17.07	8.69	1.12	1920 "	106,500	52,500	47	54	120	179
0.18	0.27	17.71	8.44	1.63	1920 "	111,500	51,000	52	60	120	187

A steel of 0.40 C, 17 Cr, 9 Mn, 2.25 W, and 0.30% Ti quenched from 2000° is stated by Wolfson and Borzidka<sup>27</sup> to show, at room temperature, 117,000 tensile, 27,500 yield, 26.5% elong., and 29.5% R.A.

Notwithstanding that the 18 Cr, 8 Mn alloy is not completely austenitic, but, at room temperature, contains both ferrite and austenite, such a steel, with 0.07% C, in the quenched condition shows about 100,000 tensile, 50,000 yield, 45% elong., 60% R.A., 120 Izod, and approaches 18 Cr, 8 Ni in workability and corrosion resistance. The corrosion resistance is not quite so good in some corrosive media, but it will serve for some purposes.

In spite of the obvious cost saving in the replacement of Ni by Mn, little use has been made here of such steels. Interest abroad is keen and future development may be expected here, when the possibilities and limitations in the application of the steels for corrosion-resistant uses are better known.

Thum<sup>28</sup> comments upon foreign use of the Cr-Mn and Cr-Mn-Ni or Cr-Mn-Ni-Cu alloys. There are so many fine points about the C-Ni-Cr alloys still to be worked out and clarified that it is no wonder that knowledge of equally fine points about these alloys, in which Mn substitutes for some or all of the Ni, is still fragmentary.

**Cobalt in Stainless.** Krivobok<sup>29</sup> suggests adding both 3% Mo and 2% Co to a 0.10 C, 18 Cr, 8 Ni alloy for heat-resistant service, citing much longer life in "time to rupture" tests at 1500° under a load of 5000 lb. per sq. in. than for plain 18 : 8. He mentions that Si and Al additions may also be made.

**More Highly Alloyed Compositions.** The 18:8 analysis is not the only combination of Cr and Ni that will give an austenitic stainless steel. For example, 20 Cr, 6% Ni; 16 Cr, 10% Ni; and 9 Cr, 22% Ni are austenitic (see Fig. 4). In order to ensure corrosion resistance of a very high order, it is necessary to hold the Cr high. To accomplish

this, around 18% Cr is required, and, with that Cr content, the least Ni that will make the austenite reasonably stable under usual heating and cooling conditions is 8%. For resistance to scaling at high temperatures, even more Cr is advisable. The alloys with higher Ni are more sluggish in carbide precipitation; so in the cast "heat-resistant" alloys in which the C may run, for ease in casting, from 0.20 to 0.40% or higher, both the Cr and Ni may be raised, 25 Cr, 12% Ni being a widely used type. Other examples are 20 Cr, 25% Ni and 25 Cr, 20% Ni. All these have low thermal conductivity, must be heated slowly, and for maximum toughness all are rapidly cooled from about 2000–2100°.

As was mentioned in the previous chapter, these austenitic alloys have much better strength at elevated temperatures than do the plain Cr steels of equal Cr contents. Since these austenitic alloys are not susceptible to heat treatment, save as they have a similar undesirable tendency toward carbide precipitation to that shown by 18 : 8, they need not be further discussed here.

All these high Cr steels are useful for resistance to scaling, to atmospheric corrosion, and to corrosion by oxidizing solutions, but are not resistant to all corrosive conditions. The Book of Stainless Steels<sup>12</sup> summarizes much of the information on their corrosion resistance.

**Relative Tonnage.** The listing of some unusual analyses should not lead the reader to conclude that they are in wide use. Lippert<sup>30</sup> has estimated the production of the various stainless and rustless alloys in 1936 and 1937. About 45% of the total production of wrought stainless alloys was of 18 : 8, in which of course part of the tonnage had small stabilizing alloy additions; another 45% was of the plain Cr rustless types, without appreciable Ni (about equally divided, between 12–14% Cr and 16–18% Cr); about 5% of wrought 25 Cr, 12% Ni; and 5% of "all other" alloys. In castings, the distribution was approximately as follows:

Percentage of Tonnage	COMPOSITION	
	% Cr	% Ni
18	25	12
14	18	8
12	16	36
6	15	60
3	19	38
Under 1%	12–30	None to a few %

The remainder, unclassified, is thought to be largely complex analyses for heat-resisting purposes.

Lippert comments on the use of 1.50–2.00% Mn in plain or Mo-bearing 18 : 8 for improved hot workability and weldability, and on the rising foreign use of 18 Cr, 8% Mn for corrosion resistance, and of 16 Cr, 16% Mn for deep drawing in making food-handling equipment. He mentions that there is some use of 7 Cr, 25 to 40% Mn, with low C, where a non-magnetic, highly machinable, material is required and high corrosion resistance is not necessary.

### BIBLIOGRAPHY

1. FRENCH, H. J., and J. W. SANDS (editors): "Nickel Alloy Steels." New York, International Nickel Co., Inc., 1934, 7 sections.
2. "The Alloys of Iron and Nickel." New York, McGraw-Hill Book Co., Inc. To be published.
3. HALL, J. H.: Austenitic Manganese Steel Castings, *A.S.T.M.-A.F.A.*, "Symposium on Steel Castings." 1932, pp. 200–214.
4. PAYNE, B. H.: Welding 11–14% Manganese Steel, *Welding Engineer*, v. 22, July, 1937, pp. 24–26.
5. SPRAGEN, W., and G. E. CLAUSSEN: Welding Malleable Cast Iron—A Review of the Literature to December 1, 1937, *Welding J.* (New York), v. 16, June, 1937, pp. 10–14 Suppl.
6. HALL, J. H.: Fe–Mn–Ni–C: A New Manganese Steel, *Metal Progress*, v. 20, November, 1931, pp. 69–72.
7. COUSANS, F.: Production and Properties of Steel Castings, *Foundry Trade J.*, v. 58, 1938, pp. 145–148, 150.
8. CLARK, D., and J. COUTTS: The Production and Application of Manganese Steel in Australia, *Trans. Am. Foundrymen's Assoc.*, v. 40, 1932, pp. 29–46.
9. KRÄINER, H.: Statische und dynamische Zugversuche an austenitischen Manganstählen (Static and Dynamic Tensile Tests on Austenitic Manganese Steels) *Archiv f. Eisenhüttenw.*, v. 11, 1937, pp. 279–282.
10. CAZAUD, R., and L. PERSOZ: "La Fatigue des Métaux." Paris, Dunod, 1937, 160 pages (Caulle, see pp. 83–185).
11. MONYPENNY, J. H. G.: "Stainless Iron and Steel." New York, J. Wiley & Sons, Inc., 2nd ed., 1931, 575 pages.
12. THUM, E. E. (Editor). "The Book of Stainless Steels." Cleveland, Am. Soc. Metals, 2nd ed., 1935, 787 pages.
13. "The Alloys of Iron and Chromium" (Vol. 2). New York, McGraw-Hill Book Co. To be published.
14. KRIVOBOK, V. N., and M. A. GROSSMANN: A Study of the Iron-chromium-carbon Constitutional Diagram, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 760–807.
15. ABORN, R. H., and E. C. BAIN: Nature of the Nickel-Chromium Rustless Steels, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 837–893.
16. KRIVOBOK, V. N., and R. A. LINCOLN: Austenitic Stainless Alloys, *Trans. Am. Soc. Metals*, v. 25, 1937, pp. 637–689.

17. THUM, E. E. (with comments by W. L. SUTTON): Aluminum Alloys vs. Stainless Steels for Aircraft, *Metal Progress*, v. 31, January, 1937, pp. 47-52.
- 17a. SMITH, H. A.: Pit Corrosion of Stainless Steel. *Metal Progress*, v. 33, June, 1938, pp. 596-600.
18. GRIMSHAW, L. C.: Manganese and Copper Additions to "18 and 8"—Armstrong Metal, *Metal & Alloys*, v. 5, 1934, pp. 264-266.
19. KINZEL, A. B.: Chromium and Manganese in Structural and Stainless, *Metal Progress*, v. 29, March, 1936, pp. 46-51, 74.
20. GÜNZBURG, I. S., N. A. ALEKSANDROVA, and L. S. GELDERMANN: Eigenschaften von nichtrostenden Chrom-Mangan- und Chrom-Nickel-Mangan-Stählen (Properties of Stainless Chromium-Manganese and Chromium-Nickel-Manganese Steels), *Archiv f. Eisenhüttenw.*, v. 8, 1934, pp. 121-123.
21. BECKET, F. M.: Chromium-Manganese Steels, *Yearbook, Am. Iron Steel Inst.*, v. 20, 1930, pp. 173-194.
22. BURGESS, C. O., and W. D. FORGENT: Constitution of Iron-Chromium-Manganese Alloys, *Metals Technology*, v. 5, April, 1938, *Am. Inst. Min. Met. Engrs.*, Tech. Publ. 911, 22 p.
23. BRUHL, F.: Gefüge und Eigenschaften von Chrom-Mangan-Stählen mit gehalten bis 1% C, 15% Mn und 30% Cr (Structure and Properties of Chromium-Manganese Steels with 1% C, 15% Mn and 30% Cr), *Archiv f. Eisenhüttenw.*, v. 10, 1936, pp. 243-255. See also KLUKE, R.: Einfluss von Manganzusätzen bis 20% auf Gefüge und Eigenschaften von Gusslegierungen mit 30% Cr (Influence of Mn additions up to 20% on the Structure and Properties of Cast Alloys with 30% Cr), *Ibid.*, v. 11, 1938, pp. 615-618.
24. SCHMIDT, M., and H. LEGAT: Hitze beständige Chrom-Mangan-Stähle (Heat Resistant Chromium-Manganese Steels) *Archiv f. Eisenhüttenw.*, v. 10, 1936, pp. 297-306.
25. HOUARDY, H., and G. RIEDRICH: Article in *Metallwirtschaft*, v. 16, December, 17, 1937, p. 1329, Abstract in *Metallurgist* (supplement to *The Engineer*), April 29, 1938, pp. 123-125.
26. LEGAT, H.: Beitrag zur Kenntnis der austenitischen Chrom-Mangan-Stähle (Austenitic Chromium-Manganese Steels) *Archiv f. Eisenhüttenw.*, v. 11, 1938, pp. 337-341.
27. WOLFSON, S. I., and A. M. BORZIDKA: Short Time Creep Characteristics of Stainless Steels, *Metals & Alloys*, v. 8, 1937, pp. 294-296.
28. THUM, E. E.: Diversification of Stainless for Special Purposes, *Metal Progress*, v. 32, October, 1937, pp. 597-599, 602-604, 608.
29. KRIVOBOK, V. N.: Alloy and Manufactures, U. S. Patents 2,117,732 and 2,117,733, May 3, 1938.
30. LIPPERT, T. W.: Stainless, *Iron Age*, v. 141, May 5, 1938, pp. 28-34.
31. STRAUSS, B., H. SCHOTTKY, and J. HINÜBER: Die Carbidausscheidung beim Glühen von nichtrostendem unmagnetischem Chromnickelstahl (Carbide Separation during Annealing of Stainless Non-Magnetic Chromium-Nickel Steel) *Zeit. anorg. allgemeine Chem.*, v. 188, 1930, pp. 309-324.
32. FRANKS, R.: Effects of Special Alloy Additions to Stainless Steels, *Am. Soc. Metals*, Preprint 6, October, 1938, 17 pages.
33. HOYT, S. L. and M. A. SCHEIL: Stress-corrosion Cracking in Austenitic Steels, *Ibid.*, Preprint 7, October, 1938, 29 pages.
34. DUMA, J. A.: Influence of Aluminum on Cast Corrosion Resisting Steel, *Ibid.*, Preprint 10, October, 1938, 23 pages.

## CHAPTER 21

### CARBON TOOL STEELS

**Contrast between the Requirements for Tool and Other Steels.** The requirements for tool and die steels are quite different from those for the steels so far discussed. Even the plain C tool steels definitely fall into the "special-purpose" class.

The C and alloy steels used for constructional purposes, e.g., engine parts and the like, may be controlled as to composition, grain size, and heat treatment, to produce toughness as the primary property, coupled with whatever level of hardness and strength can be obtained consistent with that toughness. This holds for both wrought and cast steels. When wear resistance is needed, a hard carburized or nitrided case may be produced over a tough core. To secure this toughness the constructional steels normally make use of the C range 0.10 to 0.40%. When these steels are quenched and tempered the tempering is done at high temperatures to produce sorbitic structures.

In order to secure wear resistance, rail steels are made on a higher C level, approximately eutectoid, which, without heat treatment, gives a moderately fine pearlitic structure, and fair hardness but a low order of toughness.

In spring steels the main requirement is a high yield strength, so as to produce the maximum elastic range. Toughness has to be sacrificed. The C range is relatively high and tempering after quenching is at a low temperature, to produce a troostitic structure.

**Requirements for Tool Steels.** The primary requirements in tool steels are hardness and wear resistance. Toughness of a sort is needed, but it is possible to obtain only a low order of toughness without sacrificing the necessary hardness.

In many types of tools it is necessary to have free carbides present, that is, the tool steels are very often hypereutectoid. Their heat treatment is aimed to produce either a martensitic matrix or one tempered only to troostite. As in cutting tools, the hardness and wear resistance required in wire drawing dies is chiefly produced by the presence of very hard free carbides in a relatively hard, but not too brittle, matrix.

In the operation of the average metal cutting tool, the tool tip becomes very hot through friction, so a steel must be used that is stable against softening at operating temperatures.

Plain C steel tools are used in woodworking and for taking very light cuts on metals that are not too hard but, when the service becomes severe enough to raise the tool temperature notably, alloy steels are resorted to.

**Cutting vs. Battering Tools.** The ferrite strengtheners Ni and Si are used in only a very few types of tool steel for cutting tools, but enter some of the battering tools where a higher degree of toughness is required, and the steels for hot forging dies may contain Ni or Cu. The battering tools, like rivet sets and punches, are likely to be held much lower in C than the cutting tools. Manganese, both as a ferrite strengthener and a mild carbide former, is used as the sole or the main alloying element in the so-called, non-deforming tool steels, but in most types of tool steel Mn is held to a much lower limit than is the rule in constructional steels.

A ferrite-strengthening element, Co, too expensive to use in ordinary steels and in them not accomplishing much that is not equally well accomplished by cheaper elements, is introduced into some high-speed tool steels.

**Stable Carbides.** The carbide-forming elements most commonly used are Cr, V, Mo, and W.

Cementite containing any of these carbides is harder than plain  $\text{Fe}_3\text{C}$ . Chromium carbide, though very hard, is less resistant to high temperature than the others. The stability of the carbides at high temperatures seems to increase in the order listed. The greatest increase in stability is brought about when the proportion of alloying elements to C is high enough to induce the formation of double carbides in which six metal atoms, instead of three as in cementite, combine with one C atom. Such double carbides occur in the high-speed steels.

**Limitations in Heat Treatment of Tool Steel.** The heat treatment of C steel of the lower C ranges—say, under 0.60% C—is concerned mainly with the meeting of definite physical test results through quenching and toughening. The metallurgist usually has wide latitude in the selection of C content, hardening temperature, rate of cooling, cooling medium, etc., and, furthermore, can vary the possible toughening temperature within  $500^\circ$  or more. Thus, for identical sections, it may be possible, with a 0.35% C steel, to use water, oil, or air quench, and to vary the draw temperature to produce the required results. Or it may be possible to water quench a 0.30% C steel, oil quench a 0.40% C steel, or air quench a 0.50% C steel, and, with or without changing the draw temperature, again to obtain equivalent results. The metallurgist can often use a quenching rate under that of the critical rate of cooling and yet produce satisfactory commercial test results. Inas-

much as most machining of the lower C heat-treated steels is done after heat treatment, he need not use precise care in furnace atmospheres. He usually analyzes the heating phase in terms of temperature and saturation in relation to initial structure, rather than in relation to rate of heating. With this range of carbons he has at his command many variables which he can adjust in order to give the final results desired, and he usually has several alternate methods available to produce those results.

The higher C or tool-steel analyses, however, present an entirely different picture, and stringent limitations appear on all sides. Thus, with most tools, the C content is determined within narrow limits by precedent based upon experience; the heating rate becomes important owing to the greater sensitivity of increasing C content; the hardening temperature must be held within close limits; the cooling rate must equal that of the critical rate of cooling because the purpose and use of tools require maximum "hardness"; the permissible temperature range for tempering or "letting-down" is reduced to a minimum and also is established; and, finally, the atmosphere and heating conditions must be correct inasmuch as little or no work is done upon the tool after hardening. The correct heat treatment of tool steel and tools, therefore, becomes an art in which the human element is a controlling factor.

**Grade.** The problem of selecting a proper grade of steel in relation to the work required is one hitherto commonly met by the steel manufacturer alone. Until recently he has recommended this or that steel for a given requirement, depending more or less upon his general knowledge of the purpose for which the tool is to be used, and upon the experience of his customers in the past.

For the aid and information of his customers, the steel maker usually groups his tool-steel products into various "grades" and "temper." The former term refers to the "quality" of the steel, according to the class of raw material which has been used, together with the skill and care taken in producing the finished material. The highest grades should be used for tools operating under severe working conditions, or upon which a large labor cost is to be placed, such as are found in expensive dies, milling cutters, taps, etc. For very cheap tools, a steel of lower quality might be in order. All factors must be considered in the selection of the proper and most economical steel, which will give the greatest efficiency in all senses of the word. With this in mind, the following classification of the relation of quality to use is given:

1. Finest tools and dies: expense for material the smallest item entering into the cost and upkeep of the finished tool.
2. Finishing tools for lathe and planer work; special taps, reamers,

# APPROXIMATE CARBON CONTENT FOR ORDINARY TOOLS

Carbon Per Cent	Tools
1.50	Tools requiring extreme hardness. For turning chilled-rolls and tempered gun-forgings. Roll corrugating.
1.40	Hard lathe work generally. Chilled-roll turning. Corrugating. Graver tools. Brass-working tools.
1.30	General lathe, slotter, and planer tools. Razors. Drawing dies. Mandrels, granite points, scale pivots, bush hammers, peen-hammers. Ball-races. Files. Trimming dies. Cutting dies.
1.20	Twist drills. Small taps. Screw dies, threading dies. Edge tools generally. Cutlery. Cold stamping dies, leather-cutting dies, cloth dies, glove dies. Nail dies, jewelers' rolls and dies.
1.10	Milling cutters and circular cutters of all descriptions. Wood-working tools, forming tools, saws, mill picks, axes. Small punches. Taps. Cup and cone steel. Small springs. Anvils.
1.00	Reamers, drifts, broaches. Large milling cutters, saw swages. Springs. Mining drills, channeling drills. Large cutting and trimming dies.
0.90	Hand chisels, punches. Drop dies for cold work, small shear knives. Chipping chisels. Cutting and blanking punches and dies.
0.80	Large shear knives, chisels, hammers, sledges, track chisels. Cold sets, forging dies, hammer dies, boiler-makers' tools. Vise-jaws. Oil-well bits and jars. Masons' tools.
0.70	Smith shop tools, track tools, cupping tools, hot sets. Set screws.
0.60	Hot work and battering tools generally. Bolt and rivet headers. Hot drop forging dies. Rivet sets. Flatteners, fullers, wedges.
0.50	Machinery parts. Track bolt dies where water is continually running on dies (hot work).



milling cutters, and similar tools requiring a high-grade steel; wood-working and corrugating tools.

3. Tools for general purposes.

4. Tools for ordinary purposes, such as chisels, smith and boiler-shop work, etc.

5. Tools for rough or heavy work.

Expressed in a different way, the choice of a grade of tool steel depends upon three factors:

1. The precision of the work required of the tool.

2. The relative cost of the steel in comparison with the labor involved in the manufacture of the tool.

3. The life of the finished tool and its relation to the cost of production.

The grades are listed, at current base prices,<sup>1</sup> as follows, in cents per pound: "Regular," 14; "extra," 18; and "special," 22.

**Temper.** Carbon tool steels are further classified by "temper." In tool-steel sales' parlance, this refers to the percentage of C in the steel and may be denoted by figures or letters. Such classifications generally refer to a 10-point C limit: thus, No. 7 temper may refer to 0.65 to 0.75% C, or it may represent whatever C content the individual company has arbitrarily selected. In this connection it should be noted that this "temper" does not refer to, and should not be confused with, the word temper as indicating the operation of "letting down" the steel after hardening.

General recommendations for the proper C content to use for various tools are given in the table on p. 423; these, however, must not be regarded as absolute, for much will depend upon the grade of steel and upon the exact use of the tool.

**Navy Specifications.** The United States Navy specifies the following straight C tool steel for its general requirements:

Class	I	II	III	IV
C.....	1.35-1.20	1.15-1.05	0.90-0.81	0.80-0.71
Mn.....	0.35-0.15	0.35-0.15	0.35-0.15	0.35-0.15
P.....	0.015 max.	0.015 max.	0.02 max.	0.02 max.
S.....	0.025 max.	0.025 max.	0.03 max.	0.03 max.
Si.....	0.40-0.10	0.40-0.10	0.40-0.10	0.40-0.10

*Class I.* Lathe and planer tools, drills, taps, reamers, screw-cutting dies; taps and tools requiring keen cutting edge combined with great hardness.

*Class II.* Milling cutters, mandrels, trimmer dies, threading dies, and general machine-shop tools requiring keen cutting edge combined with hardness.

*Class III.* Pneumatic chisels, punches, shear-blades, etc., and in general tools requiring hard surface with considerable tenacity.

*Class IV.* Rivet sets, hammers, cupping tools, smith tools, hot-drop forge dies, etc.; tools requiring great toughness combined with necessary hardness.

The Navy Department also maintains the requirements as to grade by requiring a steel which will stand rehardening a specified number of times without cracking.

**General Properties.** The following table shows the relative toughness and hardness of tool steel of the different C contents:

Carbon

Per Cent

- |      |  |
|------|--|
| 0.50 | Toughness only.  |
| 0.60 | Great toughness with properties suitable for hardening and tempering.        |
| 0.70 | Excellent toughness, but with cutting edge.                                  |
| 0.80 | Tough tool steel, withstanding shocks, etc.                                  |
| 0.90 | Good cutting edge but with toughness an important factor.                    |
| 1.00 | Toughness and cutting edge about equal.                                      |
| 1.20 | Great hardness combined with some toughness.                                 |
| 1.30 | Great hardness in cutting edge. Toughness slight factor.                     |
| 1.40 | Extreme hardness in cutting edge first requirement. Toughness slight factor. |

Some metallurgists consider that it is safer to select a too-hard steel and draw the temper at a higher temperature than to choose a too-soft steel with a view to increasing its hardness by a weaker tempering. Opposed to this is the fact that the higher the C content the more the care which will be required in the hardening operation, and the more brittle the core of the hardened tool.

**Manganese in Water-hardening Tool Steel.** The percentage of Mn in C tool steels to be water hardened is of considerable importance on account of its effect in deoxidizing the steel during melting, the effect of the carbide, and the tendency toward increase in sensitivity during quenching. Earlier practice tended toward keeping the Mn as low as possible, with a very definite decrease as the C content was raised. Thus, a C tool steel with 1% C and the Mn as high as 0.50 or 0.60% would exhibit a tendency to crack upon water quenching, and in general practice a very low Mn content was deemed necessary. This

resulted in specifications calling for 0.40% Mn with 0.60% C, and decreasing to 0.20% Mn with 1.20% C. The present tendency, however, is toward a slightly higher Mn content, but this is correlated with the percentage of C so that the Mn content with 0.90% C is placed at about 0.35%, and with 1.20% C at about 0.25 to 0.30%. Whether this additional introduction of Mn has more effect as an alloy or as a de-oxidizing and fluxing agent for impurities in the molten bath is still an open question. The evidence tends to show, however, that there is often much more satisfactory hardening when the Mn is raised a few points—as above indicated—and the tendency is to control the Mn content very closely.

Manganese is raised and C lowered in some oil-hardening tool steels, as will be discussed later.

**Physical Tests and Machinability.** Inasmuch as nearly all tools are used in a very hard condition, the question of static strength and similar physical test properties is given little consideration. In the full-annealed condition, average properties are: 90,000–100,000 tensile, 30,000–40,000 yield, 15 to 20% elong.

The Brinell hardness of properly annealed tool steel for average machine work should be between 156 and 174; this usually indicates a spheroidized structure with excellent machinability. Brinells of 187 to, say, 212 indicate a fine-grain, lamellar pearlite structure, with fair machinability. Higher Brinells, especially in larger sections, indicate improper or insufficient annealing with poor machinability.

As with most rules, however, there are exceptions, and such may be noted for small parts requiring fine machining, threading, and work done on automatic screw machines, as in such pieces softness leads to tearing. There are numerous cases of this kind in which Brinells of 225 to 250 are desirable and specified.

The procedure for spheroidizing was discussed in Chap. IV, Vol. I. Pearlitic structures generally are characteristic of annealing at temperatures in excess of 1400–1450°.

**Heat Treatment.** The heat treatment of high-C steel involves a consideration of:

- (a) Initial structure.
- (b) Rate and time of heating.
- (c) Temperature of heating.
- (d) Rate of cooling.
- (e) Tempering.

**Initial Structure.** The initial structure in C steels, as presented in the tool or die to be hardened, usually offers but a small part of the

difficulties common to the lower C structural steels. This situation is largely due to the fact that competitive efforts on the part of those who supply tool steel are based on quality-production costs rather than on quantity-production costs. Such production necessitates the best of raw materials; more expensive melting processes, such as the use of electric furnaces; exacting metallurgical supervision as required by an inherently sensitive steel; closer inspection; and higher-grade labor with a degree of pride in the product. The keen competition for economic survival—to say nothing of supremacy—automatically and in-



FIG. 1.—Overheated 1.05% C steel, quenched in water from 1450° and annealed at 1440°.



FIG. 2.—Improperly annealed tool steel.  
× 120. (Green.)

tensively enforces quality production. Such conditions result in a so-called buyer's market; and, in such a market, quality in which initial structure is a vital factor, is preeminent.

Aside from a decarburized skin, or "bark," which should be allowed for in machining, and dirty steel caused by poor melting methods, the most common type of initial structure causing future trouble is large grain size, either initially impressed upon the steel through excessive rolling temperatures and not removed in the general annealing process, or caused by too high annealing temperatures. These conditions are illustrated in 1.10% C steel by the micrographs of Figs. 1 and 2, respec-

tively. The steel of Fig. 1 had been water quenched from  $1450^{\circ}$  in the usual course of hardening and then annealed at  $1440^{\circ}$ . With such conditions it is evident that an air quenching or even oil quenching above the Ac<sub>m</sub> temperature must first be given in order to break up the initial structure. This may be followed by a spheroidizing treatment, outlined in Chap. IV, Vol. I. As a short cut to the long anneal when occasional pieces only are involved, an oscillating anneal may be used, that is, heating at the spheroidizing temperature, cooling to a black heat, and repeating. Such practice tends to eat away the cementite from the network in the heating, and to set it by the air cooling. Such methods, in cases in which small mass effects are involved, tend considerably to reduce the total time required.

**Rate and Time of Heating.** With reasonable initial structure, preferably spheroidized, the rate of heating is indicated by the general shape, dimensions, and mass of the piece in relation to the C content. Due care must be used to bring up to temperature the whole of the piece at the same time. Overheating one part, as against underheating an adjacent but heavier section, will result in the setting up of stresses which may result disastrously for the steel upon cooling. The higher the C content the more sensitive is the steel to non-uniform heating. Spalled corners, checks, cracks, and warping are usually synonymous with lack of proper heating.

**Temperature of Heating.** The temperature range for heating, for hardening of the hypereutectoid steels is far narrower than for the lower C structural steels. The tendency toward grain growth usually becomes apparent when temperatures over  $1450^{\circ}$  are used for the small sections common to most tool and die purposes. That temperature must be used which will equalize the effect of mass as against the critical rate of cooling, and it is very questionable whether any further increase in temperature will increase further the hardness of the steel for commercial purposes.

**Coarsening in Heating Tool Steel.** This is not strictly true with the modern controlled grain-size steels, which, though not hardening so deeply from the usual quenching temperature as do those more prone to coarsening, will stand higher quenching temperatures. A higher quenching temperature, as Digges and Jordan<sup>2</sup> have shown, lowers the critical cooling rate and thus promotes deeper hardening. Thus coining dies that are used under heavy pressure and need strength at the core may be quenched from  $1500^{\circ}$ , or even  $1600^{\circ}$ , when made from fine-grained steel.

The initial condition of the steel prior to heating for quenching affects the depth of hardening. If the steel is spheroidized before hard-

ening, it hardens less deeply than if non-spheroidized, unless the quenching temperature is above  $A_{cm}$  so that the free carbides are dissolved in the austenite.

The toughness of the inherently fine-grained tool steels ("tough timbre," high coarsening temperature) is higher than that of the inherently coarse-grained (low coarsening temperature) type, even though the steels are quenched at normal temperature and actually do have, from that temperature, the same grain size. The resistance to grain growth may be conferred by the addition to the melt, at the proper stage and with the melt in proper condition, of a regulated amount of Al. In the example studied by Digges and Jordan the presence of 0.013% total Al, as metal and oxide, instead of 0.004% made

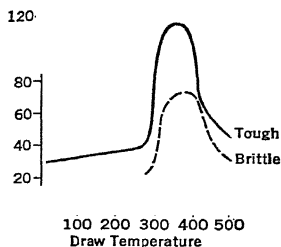


FIG. 3.—Effect of tempering temperature on torsion impact values of two C tool steels, brine quenched from 1450°. (Palmer.<sup>4</sup>)

	C %	Mn %	Si %
Tough	1.06	0.20	0.16
Brittle	1.03	0.23	0.16

the difference between a steel of high and of low coarsening temperature. Use of 0.15–0.30% V accomplishes much the same end, such C-V tool steels allowing quenching temperatures of 1600° without coarsening.

**Tempering Effects.** The difference in toughness of the low and high coarsening temperature type is most markedly depicted by torsion impact, as Luerksen and Greene<sup>3</sup> and Palmer<sup>4</sup> show (Fig. 3). The torsion impact curves show a toughness peak at a draw temperature of about 350° (1 hr.), followed by a valley at around 500°, after which it rises again. The explanation of this seems to be that quenched 1.00–1.10% C tool steel is tetragonal martensite plus some retained austenite. Tempering up to about 350° changes the tetragonal to the cubic martensite and gives stress relief, toughening the material. The toughness seems partly due to the retained austenite, for studies of length changes on tempering indicate that the retained austenite begins to transform to martensite at about 325–400°. When too much of it is lost, the toughness is lost also. At a temperature where toughness is being lost from this cause the martensite has become somewhat

tempered also, as the hardness decreases as well as the toughness. The C tool steel can be drawn back about 2 or 3 Rockwell C points from its quenched hardness to reach the peak of the toughness curve, but if softened to 60 Rockwell C the valley of the toughness curve is reached.

**Quenching Temperatures.** The following table gives the approximate temperatures for handling general tool steels; the two columns

Approximate Carbon Content, Per Cent	General Use or "Temper"	Welding Properties	Approximate Forging Temperatures, °F.	Critical Range, °F.	Approximate Hardening Temperature, °F.	
					A	B
0.60	Hot-work tools.....	Welds readily.....	1600-1800	1340-1380	1410-1460	1480
0.70	Smith tools.....	Welds readily.....	1600-1700	1340-1375	1400-1450	1470
0.80	Shears.....	Weldable.....	1600-1700	1340-1365	1390-1450	1465
0.90	Chisels.....	Weldable.....	1600-1650	1340-1360	1375-1450	1455
1.00	Reamers.....	Weldable.....	1600	1340-1360	1375-1450	1450
1.10	Milling cutters.....	Weldable.....	1550	1340-1360	1375-1430	1440
1.20	Drills.....	Requires care.....	1550	1340-1360	1375-1430	1425
1.30	General lathe work.....	Requires great care and suitable flux.....	1500	1340-1360	1375-1425	1415
1.40	Hard lathe work.....	Requires extreme care, experience, and flux.....	1500	1340-1360	1375-1425	1400

given under hardening temperatures represent the recommended practice of two well-known steel companies. The A. S. M.<sup>2</sup> gives the following, and recommends normalizing after forging and before annealing:

TEMPERATURES			
C Content $\%$	Normalizing	Annealing	Quenching
0.70-0.90	1500-1600	1380-1400	1450-1500
0.90-1.10	1550-1650	1400-1420	1400-1500
1.10 or over	1600-1700	1420-1440	1400-1475

Quenching is in brine or water. Tempering, for relieving stress, is at 300-350°, for stress relief and reduction in hardness, at 350-500°, and for reduction in hardness, 500° up.

**Rate of Cooling.** The importance of the cooling phase, particularly for the tool steels, was discussed in Chapter VI, Vol. I, "Principles of Hardening." The cooling rate must be such as will give the critical rate of cooling, for in no other way, with C tool steels, can the requisite

hardness be obtained in relation to the combination and correlation of the other factors demanded.

**Tempering.** The tempering colors used for average tools are given as follows:

- Faint yellow: Steel-engraving tools.  
Light turning tools.  
Hammer faces.  
Planing tools for steel.  
Ivory-cutting tools.  
Planing tools for iron.  
Paper-cutting knives.  
Wood-engraving tools.
- Light yellow: Milling and other circular cutters for metal.  
Bone-cutting tools.  
Scrapers for brass.  
Shear blades in general.  
Boring cutters.  
Leather-cutting dies.  
Screw dies.  
Inserted saw teeth.  
Taps.  
Rock drills.  
Chasing tools.  
Penknives.
- Straw: Dies and punches in general.  
Molding and planing cutters for hardwood.  
Reamers.  
Gouges.  
Brace bits.  
Plane irons.  
Stone-cutting tools.
- Deep Straw: Twist drills.  
Cup tools.  
Wood borers.  
Circular saws for cold metal.  
Coopers' tools.  
Augers.
- Brown: Drifts.  
Circular cutters for wood.  
Dental and surgical instruments.  
Axes and adzes.  
Saws for bone and ivory.
- Peacock: Cold sets for steel and cast iron.  
Hand chisels for steel and iron.  
Boilermakers' tools.  
Firmer chisels.  
Hacksaws.
- Purple: Molding and planer cutters for soft wood.  
Smith tools and battering tools generally.



Blue:           Screwdrivers.  
                  Saws for wood.  
                  Springs in general.

Tempering at a definite temperature for a definite time, to produce the exact Rockwell C hardness desired, is decidedly preferable to attempting to draw by color.

#### BIBLIOGRAPHY

1. *Daily Metal Trade*, September 23, 1938.
2. DIGGES, T. G., and L. JORDAN: Hardening Characteristics of One Per Cent Carbon Tool Steels, *Trans. Am. Soc. Metals*, v. 23, 1935, pp. 839-860.
3. LUERSSEN, G. V., and O. V. GREENE: Interpretation of Torsion Impact Properties of Carbon Tool Steels, *Ibid.*, v. 23, 1935, pp. 861-885.
4. PALMER, F. R., "Tool Steel Simplified." Reading, Pa., Carpenter Steel Co., 1937, 316 pages.
5. American Society for Metals: "National Metals Handbook." Cleveland, The Society, 1936 ed., page 695.

## CHAPTER 22

### LOW AND INTERMEDIATE ALLOY TOOL AND DIE STEELS

**Gradations among Tool Steels.** There are so many compositions and brands of tool steels, from C to super-high-speed, nearly a thousand brands being on the market, that at first sight it would appear impossible to group them into a reasonable number of classes. That slight gradations in one quality or another fit the steels for specific uses is evidenced by the survival of many types and modifications of types differing only minutely from each other.

The ramifications of tool steels would require a book rather than a mere chapter for anything like complete treatment. The reader is referred to Gill,<sup>1</sup> Palmer,<sup>2</sup> Knowlton,<sup>3</sup> the "A.S.M. Handbook,"<sup>4</sup> and the Alloys of Iron<sup>5</sup> series of monographs for detailed discussion. All that will be attempted here is to trace the effects of the various alloying elements in some of the more important types of tool steels in order to give an inkling of the choice among commercial types for different kinds of service and of the way other steels might be built up to meet special conditions of service or varying economic conditions as to cost of alloying element.

**Toughness vs. Hardness.** Nevertheless, certain generalizations can be made. In battering tools, toughness is very important; so these are hypoeutectoid to eutectoid, and the pure cutting tools range from eutectoid to strongly hypereutectoid, the free carbides in the latter serving somewhat as do the abrasive grains in a grinding wheel. Cutting power depends on the presence of hard carbides. If the use of the tool does not appreciably raise the temperature of its cutting edge, hypereutectoid C steel is the basis of the tool. The matrix must also be hard; normally it is slightly tempered martensite or, as toughness becomes more important, secondary troostite. The plain C steels take keener edges than do the bulk of the alloy steels.

**Toughness.** Razors, as every user knows, are relatively brittle; they are not expected to be subjected to shock or to high-temperature service. They usually run from 1.00% to 1.20% C. Files need to be almost equally hard, but the teeth meet some shock in service; so they ordinarily approximate 1.00% C. A rock drill, or a machinist's punch,

is classed more as a battering than a cutting tool; so such tools run a bit under eutectoid. Most cutting tools require a hardness when ready for service of 55–65 Rockwell C; the battering tools normally range from 45 to 55. The hardener's chief aid in evaluating the results of heat treatment and in verifying the uniformity of results is the simple hardness test, which, at the range of hardness involved usually means a file test or the diamond cone or pyramid test. When the service will permit lowered toughness, the upper hardness ranges are used in both classes. In the higher hardness ranges, with the brittleness that accompanies them, tensile testing is difficult, the preparation of test bars is costly, the standard 0.505-in. bars seldom reflect accurately the properties of the cutting edge or the toughness of the tool as a whole. The notched-bar impact figures on standard bars are so low that such tests are not very discriminatory, though Levinsohn and Dowdell<sup>6</sup> find them helpful in the study of some of the lower hardness ranges. Tension impact, i.e., a rapid tension test on unnotched bars, is almost as costly a test as a regular tension test and is little used. Torsion impact is being relied upon for evaluation of toughness by several tool steel makers and tool manufacturers.

These various means of studying toughness all tend to show that toughness increases as martensite is progressively tempered to troostite and that this general trend is modified by the retention of austenite in the quench, its preservation in low-temperature tempering and its transition to martensite and in turn to troostite. Austenite tends to toughen, martensite to embrittle. The failure of a tool is not independent of the carbides and their distribution, since the carbide particles may act as stepping stones for transmission of stress and propagation of a crack. Internal stresses and their relief also have an effect upon toughness.

**Service Evaluation of Toughness.** With all these influences affecting toughness, it is not remarkable that the different methods of measuring toughness fail to tell exactly the same story, or that the peaks of toughness and valleys of brittleness in the curves plotted against quenching temperatures, or against tempering temperatures or times, do not always come at the same ranges under the different methods of toughness testing.

The situation is somewhat like that of testing for cutting ability, where a steel or a heat treatment that is best for a lathe tool may not be by any means best for a drill, a broach may require something still different, and so on for the various varieties of tools. And when the material cut is changed, the best tool angle and the best material and treatment almost invariably change also.

Performance tests in actual service of exactly the type to which the tool is to be matched are, therefore, the best and almost the only criterion of toughness. Once satisfactory performance has been obtained, meticulous attention must be given to maintaining exactly the same conditions of quenching and tempering. And the reproduction of the heat-treatment conditions is unavailing unless the steel always responds alike.

As has been brought out in discussion of grain size (Chap. 9, Vol. I) and of the "timbre" of C tool steel (Chap. 21), steels of varying coarsening capacity do not respond alike. And, of course, variations in composition, especially in C content, are also reflected in the response. In the past the purchaser of tool steel has relied on the maker to supply, under a given brand name, material of constant response, and the maker has relied on making that brand from definite raw materials and in a definite way, without necessarily knowing just how or why he did make a uniform product. Hence not all the products were really uniform.

**Specifications and Testing *vs.* Brand Names.** The evaluation and control of coarsening capacity, by use of the various hardening and fracture tests, are helping to correct this situation.

Thus a given type of steel may be ordered to a certain chemical composition, with specified (usually fine) grain size and known coarsening characteristics (usually a high-coarsening temperature in order to favor toughness). The use of grain-size control, of a suitable depth-hardening alloy addition to make up, when necessary, for the lower depth hardening of the fine-grained steels, and perhaps a readjustment of C necessitated by change in alloy content, allow controllable gradations. With such control on the part of all tool steel makers, one may now obtain steels of the desired performance, irrespective of the source of supply, to a much greater degree than when chief reliance had to be put on brand names.

**Cost a Minor Factor.** The cost of the metal in a tool or die is ordinarily negligible in comparison with the work of making it. Hence there should be little urge to cut the base cost of the steel by choice of the cheapest alloy combination that would serve, and any alloy addition that is found to help in any degree deserves to be made, even though it may slightly raise the cost. The great number of alloy modifications of a given base type is evidence of the practicability of substituting, for part or all of one alloying element, proper amounts of one or more of a variety of other alloying elements, when such a change brings about an improvement.

These steels are ordinarily made in small melts so that there is not

the urge for standardization that comes when large open-hearth furnaces are used for melting alloy steels under large tonnage requirements.

**Ferrite-strengthened Tool and Die Steels.** The ferrite strengtheners are little used in tool steel. However, the lower C battering type tools are a sort of transition from spring steels to tool steels; they require high yield strength, a good measure of toughness, and do not need extreme keenness of cutting edges. So silico-manganese combinations form a base for such steels, with C held down to the water-quenching range, say 0.50–0.75% C, in order to get maximum yield strength. A typical composition will range from 1.50 to 2.25% Si with Mn from 0.70 to 1.75%, but not with C, Si, and Mn all on the high side.

Added to such a base, with Si and Mn suitably reduced, there may appear, singly or together, Cr or Mo from about 0.25 to about 0.50%, and V from 0.15 to 0.40%. Cr is sometimes omitted and Mo raised to 1.25%. Cr and Mo serve to add depth-hardening and oil-hardening propensities; V gives control of grain growth and added toughness; and all three tend to give more wear-resistant carbides and hence to help cutting ability. Like the Si spring steels, these steels are prone to decarburization.

The graphitic Si steels (see Chap. 4) also offer interesting possibilities for some types of dies.

**Battering Tools from Complex Steels.** For battering tools of large section where oil hardening is called for to minimize cracking, high-C, Ni-Cr steels to which may be added Mo or V, or both, are available, running 0.50–0.75% C, 0.50–1.0% Cr, and somewhere around 1.75% Ni, usually with Mo at around 0.25% or, in the lower carbons, around 0.75%. V, when used, is in the order of 0.20%. That is, these are our old friends, the rather heavily alloyed, easily hardening, steels, with C at the upper practical limit. Here the ferrite-strengthening power of Ni is utilized, along with the carbide hardness and stability of the carbide-forming elements. Chromium, V, and Mo all help in resistance to tempering. These steels, therefore, can be employed in somewhat elevated temperature service, but these steels do not have “red hardness,” i.e., will not serve as high-speed tools, though the more highly alloyed ones serve for dies for hot forging.

**Manganese-strengthened and Non-deforming Steels.** The other important class of tool steels to which some ferrite strengthening may be ascribed, is the “non-deforming” type which contain high Mn and with which oil quenching is employed to give austenite retention on quenching, in order that the dimensional changes between the machined piece ready to harden and the hardened and tempered piece,

may be at a minimum. This is required in blanking dies and punches, in some broaches, and the like, where very little or no grinding to final dimensions is desired.

The steels are hypereutectoid, with 0.85–1.00% C and 1.00 to 1.75% Mn. The Mn may be somewhat reduced and fractional percentages of Cr, W, or Mo, or some combination thereof added to give carbides of a better cutting type, and the usual 0.20% V may be included to ensure against grain coarsening at too low temperature, or other grain-size control elements may be used.

The alloy content is sufficient (1) to permit a milder oil quenching to be substituted for water hardening, thus producing a steel freer from warping, cracking, and checking or spalling; (2) to give a greater depth hardness than with a C steel quenched in water; and (3) to produce, with proper tempering, a final tool with a minimum sacrifice in hardness and yet very close in size to the original dimensions. The original basic analysis has been varied by the substitution of Cr for a part of the Mn to give greater depth hardening and surface hardness, as well as by the addition of W to give greater density and cutting properties. Typical chemical analyses, are as follows:

	U. S. Navy Grade V			
C.....	0.80–1.05	0.85–0.95	0.85–0.95	0.85–0.95
Mn.....	1.25–2.00	1.50–1.75	1.00–1.20	1.10–1.20
Cr.....	.....	.....	0.40–0.55	0.45–0.55
W.....	.....	.....	.....	0.45–0.55
Si.....	0.25–0.50	.....	.....	.....

Hardening temperatures used in practice vary between 1400 and 1500°, depending upon the shape and mass of the piece; for average work, temperatures of 1440–1460° are usually recommended. "The A.S.M. Handbook" \* suggests 1450°, oil, 325–400° draw.

Austenite is retained in large amounts through normal hardening, with the amount increasing with the quenching temperature. Upon reheating, the austenite apparently is stable up to 400°, but slightly beyond that temperature is released sharply and quite completely, with a corresponding change in mechanical dimensions. Depending upon the analysis of the steel used, the size and shape of the piece, and the purpose of the tool or die, the draw temperature will vary from about 300° or 350°—sufficient to relieve the stresses—to 450°,—at

which additional toughness is obtained as well as a decomposition of the austenite.

The dimensional changes are influenced by quenching temperature and tempering temperature and time, and to maintain the closeness of dimensions, which is the primary reason for the use of this class of steel, all the conditions of heat treatment must be kept very constant indeed.

Since the size and contour of the piece influence the final changes from the original dimensions, the original dimensions have to be worked out on a cut-and-try basis.

In order to secure hardness and get the toughness due to retained austenite, these steels are seldom drawn much above 400°. The compositions have been adjusted in this type of steel so that at a 300–400° draw the dimensional changes of quenching and tempering most nearly cancel out, and better toughness is obtained than at 450–600°, which tempering range is to be avoided.

The base price for oil-hardening tool steel is quoted <sup>s</sup> at 24 cents per lb.

**Carbide Formers in Tool Steels.** The tool steels that are aimed to be true cutting tools, in contrast to the punch-and-die type, seldom utilize ferrite-strengthening elements; instead those whose chief effect is as carbide formers are used.

For wear-resistant carbides in cutting tools at ordinary temperatures, Cr is relied upon mainly. For hardness at elevated operating temperatures, W, Mo, and V are used in combination with Cr. For keenness of edge, as in "fast finishing" steel, straight W can be used.

**Vanadium Tool Steels.** We may first consider the plain V, plain Cr, and plain-W steels. Vanadium, 0.10–0.20%, is introduced into plain C tool steels for grain-size control. The amount may rise to 0.30–0.50% in order to produce a carbide of somewhat increased cutting efficiency in a shallow hardening steel. The heat treatment of C-V steels is not materially altered from that of a C steel, save that somewhat increased depth hardening may be obtained by raising the quenching temperature somewhat, since the V permits this to be done without harmful grain coarsening.

**Chromium Tool Steels.** The desire to increase the hardenability of C tool steel and avoid danger of soft spots leads to the introduction of small amounts of Cr, usually around 0.50%, though even as little as 0.10% shows its effect. (See Chap. 7, and compare Jameson.<sup>9</sup>) The steels act much like plain C steels, but have increased wear resistance due to the entrance of Cr into the carbide. The critical, and hence the quenching, temperature is raised in proportion to the Cr content. To

increase the hardenability and wear resistance, the Cr may be raised to 0.75–1.25%. Such steels may be accompanied by 0.10–0.20% V for grain-size control. They are ordinarily water hardened. For toughness the C may be reduced as low as 0.50%, when sufficient Cr is added to give the desired hardening.

Such steels are usually low in Mn, say 0.30%, but in a variant the Mn may be raised to around 0.50%, when oil quenching becomes necessary. This illustrates the sensitivity of the tool and die steels to small changes in composition, as to alloy content. Similarly, C content needs to be carefully controlled. One firm produces two tool steels for different purposes with the same amounts of three different alloying elements, but varying in C by 0.05%. This is sufficient to throw the two steels into different groups. A steel of 0.85 C, 2% Cr, water quenched, is used for cold rolls that must have high surface hardness.

By raising the Cr to around 3.25–4%, the steels become air hardenable with C from 0.85 to 1.00% and suitable for oil quenching with it at about 0.70%. Such steels are low temperature "hot die" steels, useful when the operating temperature is not over 600°. These steels are modified by addition of W or Mo, but before discussing such analyses, a still higher Cr type should be mentioned.

**High Carbon-High Chromium Steels.** In the "high-C-high-Cr" type with 12 to 14% Cr and 1.50 to 2.50% C, large amounts of the Cr-Fe carbide  $(\text{Cr, Fe})_7\text{C}_3$  are present, for wear resistance in dies, shear blades, and the like. These steels are not ordinarily classed as tool steels, since they behave better in such service as drawing dies, lathe centers, etc., where they have to resist frictional wear, rather than in use for removal of metal as a tool.

A slightly modified air-hardening steel for automobile body dies and for blanking dies contains about 1.60 C, 12 Cr, 0.75–1.00 Mo, and 0.25–1% V. The small dimensional changes after hardening and tempering as well as the wear resistance are factors in these uses.

Wills<sup>10</sup> discusses the variations in toughness (measured by un-notched Izod specimens) in such a steel with different hardening temperatures and different drawing temperatures and times. Among these, holding for  $\frac{1}{2}$  hr. at 1800–1900° and drawing at 400–800° to about 58 Rockwell C gives high toughness. The temperature range of heating for air hardening is quite wide, and the time of holding may be quite short. Holding time of 3 hr. gave decreased toughness, however. The high-C-high-Cr steels have low thermal conductivity and demand very slow heating. They are air or oil hardening and resist tempering. Since Cr raises the critical, quenching temperatures of the order of 1800° are required. Some austenite retention on quenching is noted,



and still more is readily brought about by raising the temperature still further. On tempering at 900–1000° this austenite may be broken down with the appearance of secondary hardness, akin to that of high-speed steel. As in the Mn non-deforming steels, this austenite retention allows preservation of close to the original dimensions, by adapting the quenching and the tempering temperatures to the particular composition used. However, the steels are generally tempered around 400° for maximum hardness or around 800° for maximum toughness.

The steels are somewhat prone to decarburization, so need protective packing or the use of controlled atmospheres for hardening. The steels can be nitrided, and interesting results as to wear resistance have been secured by giving them a nitrided skin.

The base price for high-C-high-Cr steel is given <sup>s</sup> as 43 cents per lb.

Like other classes of tool and die steels, the high-C-high-Cr class is modified by various additions, among which are 0.50% Ni, 0.75–1.00% V (an amount sufficient to modify the carbides), 0.50–1.00% Mo, for increasing air hardening, and 0.50–1.00% Co, which confers some red hardness and is likely to be added when this type is used for cutting tools. Various combinations of such additions, as well as a higher Cr grade running around 17% Cr, with perhaps 1.60% C, are met. Houdremont <sup>11</sup> mentions 2.40 C, 27% Cr for drawing dies.

**Mushet Steel.** The first "high-speed" type of steel, made by Mushet in 1868, was a high, 2.00–2.40% C steel, with 5 to 8% W, with high Mn, 1.75–2.50%, high Si, around 1%, and to this combination was soon added 1.00–1.25% Cr. This was the original air-hardening tool steel. In later developments the C was lowered and Cr took the place of Mn, the C-W-Cr combination forming the basis for later developments in hot-working and high-speed steels.

**Tungsten Tool Steels, Fast-finishing Steels.** The plain W or "fast-finishing" steels form a class of tool steels of the order of an intensified C tool steel. They are high C, around 1.25%, with 3 to 6% W. The tungsten carbide is harder than cementite and more wear resistant. Tools of this steel take a keen edge. They are used chiefly for cutting tools, e.g., taps, hacksaws, and to some extent for wire drawing dies. They quench intensely hard in water and are used for much the same purposes as plain C tool steels, but will cut harder materials.

Oddly enough, in view of the use of W in heat-resisting steels, these plain C-W steels, as usually treated, are not especially resistant to tempering, nor do they exhibit red hardness. As Houdremont shows <sup>11</sup> a great increase in depth hardening may be obtained in a 1.50 C, 8% W steel by raising the quenching temperature, and when so hardened there is marked resistance to tempering. The carbide must be in solu-

tion and well diffused for these effects to show up. There is possibility of graphitization in heating such steels. They are prone to decarburization. If spheroidized on annealing, the carbides are hard to redissolve, and hardening may not be deep enough. The steels are shallow hardening anyway. In Germany a steel of 1% each of C, W, and Ni has been employed for better depth hardening.

The introduction of Cr along with W changes the W steels from shallow hardening to deep hardening, and as Cr is raised a degree of "red hardness," not met with Cr alone or W alone, is produced. Vanadium and Mo are added also, and the more complex of these steels begins to approach high-speed steels in some respects.

**Oil-hardening Tungsten Chromium Steels.** With 0.90-1.25 C, 1.00-2.50 W, and about 0.50% Cr, the steels shift from water hardening to oil hardening. They avoid the shallow-hardening troubles of the plain W steels but retain some of their ability to take keen edges. Thus this class makes good taps and reamers.

Lower C and higher Cr content, e.g., 0.45-.65 C, 1.00-2.50 W, 1.00-1.50% Cr, often with 0.20% V, give greater toughness, and this is essentially an oil-hardening chisel, punch, and battering-tool grade. It has sufficient Cr and W to begin to develop resistance to tempering and a degree of red hardness. Thus it is suitable for hot working tools where toughness is paramount, and where operating temperature is not excessive. About 0.75% Mo may be substituted for the W in this type.

**Hot-work Steels.** For hot-work die blocks a steel of 0.50 C, 0.75 Mn, 0.30 Si, 1.00 Cr, 0.40 Mo, and 0.15% V is advocated as being especially free from cracking. In order to avoid too strong an air-hardening tendency, Ni is kept out of this composition by some producers. Finkl<sup>12</sup> advocates a die-block steel of 0.50 C, 0.75 Mn, 0.75 Cr, 1.00 Ni, 2.00 Cu, and 0.25% Mo, for good depth hardening and resistance to heat checking.

The hot-work steels for service at more elevated temperatures use lower C, 0.25-0.50%, and raise the Cr 3 to 7% with the W at 3 to 11%, often with up to 0.5% V and sometimes with Si up to 1%. They approach the composition of high-speed steel diluted with Fe. Part or all of the W may be replaced by Mo. Such steels are used in dies for hot forming. A German hot-die steel<sup>11</sup> has 0.70-0.90 C, 0.50 Mo, and 3% Cr. Another steel contains 0.70 C, 0.25 Mn, 3.80 Cr, 0.80 Mo, and 0.60% V. This can be hardened in 1 in. round to 64-65 Rockwell C by cooling in an air blast from about 1850° or in oil from 1800-1850°. The steel is strongly depth hardening, 4½-in. cubes showing 54 Rockwell C in air and 62 Rockwell C in oil. The steel has strength at high

temperatures and is very resistant to tempering. When quenched to 64 Rockwell C it takes 1000° to soften it to 55 Rockwell C. Saklatwalla<sup>13</sup> suggests, for air hardening tools and dies, steel of 0.60–1.25 C, 0.50–1.50 Si, 0.10–1.00 Mn, 3.00–6.00 Cr, 0.60–3.00 Mo, and 0.30–1.50% Cu, and for high-temperature service, the same with lower C, 0.05–0.40%.

If the Cr in the Cr-W hot-work steel is around 7%, the W is held to about the same limit; with Cr in the neighborhood of 3%, W may range all the way from 1% up to say 11%, with C also grading from 0.25 to 0.60%. With this Cr content, the addition of more W raises red hardness and decreases toughness, and lower C increases toughness. Just as in high-speed steel, V adds to hardness and high-temperature stability, as well as aids toughness by its grain control.

A class of highly alloyed steels, used for hot-work dies and for severe service as molds for die castings, resemble high-speed steel diluted with Fe, i.e., with lower C and with somewhat less of all the alloying elements. Cobalt may be added, just as in high speed. Some of the steels differ from the high-speed type in having higher Cr, and in the addition of Si.

Examples are:

C	W	Mo	Cr	V	Co	Si
0.30	9.00	....	3.00	0.40	.....	....
0.40	1.00	5.75	3.50	0.75	.....	....
0.30	....	9.00	5.00	....	0.50–0.75	3.50
0.35	....	5.00	5.50	0.50	4.00	1.00

The second steel in this table is reported<sup>14</sup> as giving 56 Rockwell C after hardening from 2240° and drawing at 1100°.

If, instead of hot-die steels, we seek cutting tools to operate at high temperature, the C is raised to 0.65% or above and the alloy content is increased. At 4 Cr, 14% W we reach the high-speed steel class, which will be discussed in the next chapter.

Inserts for exhaust valve seats on Ford automobile engines are stated by Cone<sup>15</sup> to be made from high-speed steel scrap, with C raised by addition of cast iron, and with Cr added. The composition runs 1.20–1.40 C, 14.00–17.00 W, 2.50–3.50 Cr, 0.30–0.60 Si, 0.30–0.50 Mn, 1.50–2.00% Cu. As cast, it shows 50–60 Rockwell C. Cooling from 1450° to 1000° at the rate of 125°/hr. softens and toughens it to 40–46 Rockwell C.

## BIBLIOGRAPHY

1. GILL, J. P.: "Tool Steels." Cleveland, Am. Soc. Metals, 1934, 136 pages.
2. PALMER, F. R.: "Tool Steel Simplified." Reading, Pa., Carpenter Steel Co., 1937, 316 pages.
3. KNOWLTON, H. B.: "Heat Treatment, Uses, and Properties of Steel." Cleveland, American Society for Metals, 1929, 437 pages.
4. American Society for Metals: "National Metals Handbook." Cleveland, American Society for Metals, 1936 edition.
5. The Alloys of Iron Series. Published by McGraw-Hill Book Co., New York.
6. LEVINSON, N., and R. L. DOWDELL: Toughness of Tool Steels by the Charpy Impact Test., *Bull. Am. Soc. Test. Mat.*, v. 7, December, 1935, p. 4.
7. BAIN, E. C., and M. A. GROSSMAN: The Nature of Oil Hardening Non-deforming Tool Steels, *Trans. Am. Soc. Steel Treat.*, v. 10, 1926, pp. 883-897.
8. *Daily Metal Trade*, September 23, 1938.
9. JAMESON, A. S.: Cold Heading Die Life, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 313-350.
10. WILLS, W. H.: Further Study of a High Carbon High Chromium Tool Steel, *Trans. Am. Soc. Metals*, v. 25, pp. 1013-1029.
11. HOUDREMONT, E.: "Einführung in die Sonderstahlkunde." Berlin, Springer, 1935, 566 pages.
12. FINKL, W. F.: Die Block, U.S. Patent 2,104,979, January 11, 1938.
13. FINKL, W. F.: Steel Alloy, U.S. Patent, 2,104,980, January 11, 1938.
14. SAKLATAWALLA, B. D.: Alloy Steel, U.S. Patent, 2,102,283, December 14, 1937.
15. Molybdenum-Tungsten High Speed Steel Is Finding Wide Acceptance in Industry, *Steel*, v. 100, March 15, 1937, pp. 66-68, 98.
16. CONE, E. F.: Manufacture of Valve Seat Inserts for Ford Engines, *Metals & Alloys*, v. 8, 1937, pp. 89-91.

## CHAPTER 23

### HIGH-SPEED TOOL STEELS

**Requirements of High-speed Steels.** The requirements in high-speed tools are that the steel produce a cutting edge which shall not be rapidly dulled when operating at such speeds and feeds that friction heats the tool, even to redness. A certain degree of toughness is required, i.e., the tools must not be glass-brittle.

The necessary combination is found in a relatively small range of compositions, heat treated in quite definite fashion. Free carbide particles are essential; the steels must be hypereutectoid. The carbides and the matrix must be stable and hard at elevated temperatures.

**Carbides in High-speed Steel.** Carbides of this type are obtained by using a ratio of carbide-forming elements to C much higher than in constructional steels. Alloyed cementite is considered to be  $\text{Fe}_3\text{C}$  in which some of the Fe is replaced by Mn, Cr, Mo, or other carbide-forming element, but the cementite molecule is still like that of  $\text{Fe}_3\text{C}$  and the formula can be written  $(\text{Fe,Cr,Mn,Mo})_3\text{C}$ , meaning that any proportion of the metallic elements is combined with C in the proportion of three metal molecules to one C molecule.

The carbides in high-speed steels, however, are double carbides, whose composition can be expressed, in the case of a plain W or Mo steel, by the formula  $(\text{Fe}_3\text{W}_3)\text{C}$  or  $(\text{Fe}_3\text{Mo}_3)\text{C}$ , (though some evidence indicates that  $(\text{Fe}_4\text{Mo}_2)\text{C}$  might be the correct formula). At any rate, the alloyed carbide can be denoted by  $(\text{Fe,W,Mo,Cr,V})_6\text{C}$ . When six such metal atoms combine with one C atom, the "double" carbide formed is far more temperature resistant than is cementite carrying other carbide-forming elements beside Fe.

**The Equivalent Iron, Double-carbide Diagram.** There is an Fe, double-carbide diagram analogous to the Fe-C or Fe-cementite diagram.

Grossmann and Bain<sup>1</sup> point out that this fact allows consideration of high-speed steel as entirely analogous to ordinary steel. They plot the diagram of Fig. 1 for Fe and double carbide, in which the point S represents the eutectoid, and the line *SE* the Acm line. Note the differences in temperature from those of the Fe-C diagram. The useful high-speed steels contain carbide corresponding to compositions from

about half way up on the *SE* line, up almost to *E*. In order to exceed *Acm* sufficiently to take these difficultly soluble carbides into solution, *Acm* (*SE*) must be considerably exceeded, so high-speed steel has to be heated to 2200–2400° to harden properly.

The location of the line *EJ* indicates a temperature which cannot be exceeded without the steel's beginning to melt, when a "burnt" steel

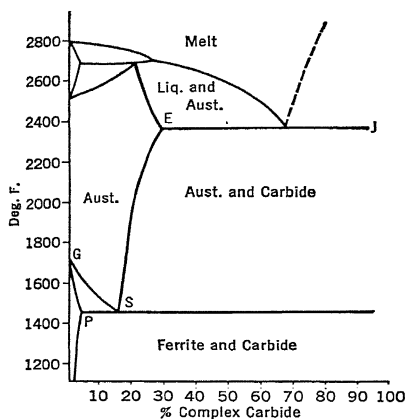


FIG. 1.—The pseudo-binary constitution diagram of Fe-Fe<sub>4</sub>W<sub>2</sub>C. (Grossmann and Bain.<sup>1</sup>)

This is not a true constitution diagram but serves as an approximate constitution chart.

results. The location of *EJ* will vary somewhat with the composition of the steel. It can be a trifle higher than the temperature shown.

The *Ac*<sub>1</sub> transformation occurs at about 1450° instead of 1335° for the C steels, so we may temper high-speed steel higher than C steel without getting into the austenitic field.

The eutectoid in this diagram corresponds to that of a simple steel containing about 26 W, 0.55% C. The high-speed steels can be considered as steels of that family, in which the other carbide formers Cr, V, and Mo can take the place of W, and in which the Fe can hold Co in solution as a ferrite strengthener.

**Replacement.** If we assume that 1% W in the double carbide may be replaced by 2% Cr, by 1/8% V or by 4/7% Mo,<sup>2</sup> we find that most commercial high-speed steels, such as 18 W, 4 Cr, 1% V; (18 : 4 : 1) 1.5 W, 9 Mo, 4 Cr, 1% V, or 14 W, 3.5 Cr, 2% V, figure out, on these approximate equivalents, to somewhere around 22 to 30% W. Some foreign steels go as high as 26 W, 6 Cr, 0.75% V, but, on this basis of equivalents, the high-speed steels do not vary from each other more than the members of any other class of steels, in spite of the apparent differences in composition.

**Sluggishness.** With so much alloy it would be expected that the high-speed steels should have very sluggish austenite, that they should harden on relatively slow cooling, and that it should be easy to retain considerable proportions of austenite. All these expectations are realized, the steels harden in oil or air, and retain much austenite;  $A_r'$  is entirely absent;  $A_r''$  occurs at 350–400°.

Both the martensite and the austenite are sluggish. The martensite tempers very slowly indeed and the austenite does not break down to martensite until the range 700–1100° is reached. On tempering the hardened steel, slight softening occurs up to about 700°, but from there on up to 1000 or 1100° the hardness increases (secondary hardening) due to transformation of retained austenite to martensite.

**Retained Austenite.** In order to complete the decomposition of

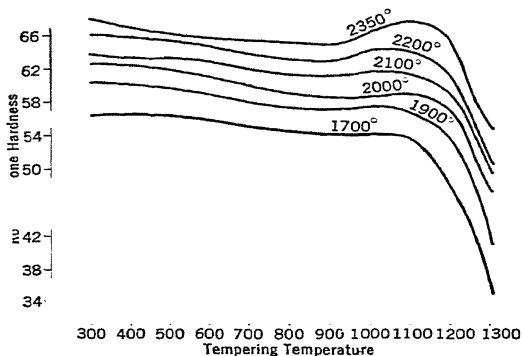


FIG. 2.—Secondary hardening in a high-speed tool steel, quenched as shown and drawn 1 hr. at temperatures shown. (Grossmann and Bain.<sup>1</sup>)

austenite and simultaneously to temper the lately born martensite a trifle, reheating may be carried, for maximum hardness, to around 1100°. The maximum toughness, however, as judged by torsion impact, occurs at about 900°, where the original martensite has been slightly tempered, some retained austenite has transformed, but some austenite is retained. Since hardness is so essential, this temperature is usually exceeded, and secondary hardening produced.

**Secondary Hardening.** In common with other alloy steels, increasing the temperature of heating before quenching increases the amount of retained austenite and also the ability for secondary hardening. (See Fig. 2.) This strong secondary hardening of high-speed steel is probably largely due to V.

**Grain Growth.** At the very elevated temperatures of heating, grain growth is rapid, even in steels containing V. Toughness is sacrificed as the grain size increases, so a compromise is required, i.e., the heating for quenching is at a high temperature but for a very short time.

**Precautions in Heating.** These steels have low thermal conductivity and hence may not be put in a furnace running at temperature for quenching, without preheating.

The ability of the steel to form sufficiently homogeneous austenite on heating is affected by the original size and distribution of the car-

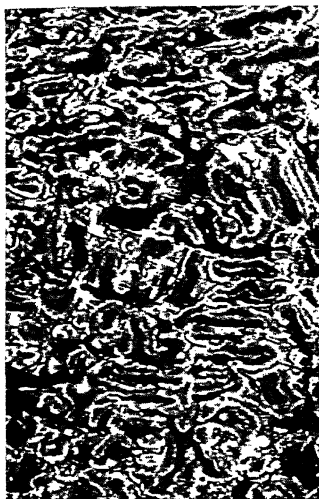


FIG. 3.—Cellular structure near outside of ingot.



FIG. 4.—Envelope structure at inner part of ingot.

5%  $\text{HNO}_3$  etch,  $\times 175$ . (MacFarland.) Varying structure in high-speed steel ingot.

bides, so the annealing to put the steel in condition for machining prior to heat treatment must take this into consideration. Not all the carbide is dissolved on heating for quenching, and the residual carbide should be as fine and as uniformly disseminated as possible.

**Avoidance of Network Structure.** When the ingot freezes the carbide forms a network, and this network must be broken up by hot work and by annealing. This is more effectively done on small ingots than on large ones, so the quality of high-speed steel depends upon ingot size and hot-working practice, quite as much as composition.



Cellular and network structures in an as-cast, unannealed high-speed ingot are shown in Figs. 3 and 4.

The elimination of the massive carbide skeletons is vital to the production of the best high-speed steel structure, and carbide out of solution, in any form other than that of small spheroids, is invariably found to be detrimental. Whereas in ordinary C steels the excess Fe carbide is readily taken into solution on heating above Acm, in high-speed steels it is found that the complex carbides in massive form, whether present as skeletons, envelopes, or laminations, are extremely refractory and almost impossible of solution by any heating process alone. The ordinary anneal has practically no influence; and the usual hardening treatment, owing to the necessity of its being carried out as quickly as possible, is entirely insufficient to break up the massive carbide structure. As has been shown by Andrew and Green,<sup>2</sup> even three successive heatings for six-hour periods at temperatures of 2100–2200°, each followed by slow cooling in the furnace, are insufficient to break up the massive carbide skeletons which exist in the original cast structure; for, although they are thinner and less massive after each successive annealing, they still remain very sharply defined.

The only method for obtaining the desired solution or emulsification of the massive carbide constituents is by hot working, such as forging and rolling, in combination with annealing.

**Forging.** Forging should be carried on from 2150° down but should include final forging at least as low as 1800° in order to get a well-broken-down structure, but not below 1600°.

The amount of reduction in forging and rolling which the metal undergoes between the cast ingot and the finished bar, and the necessity of having this working take effect to the center of the ingot or billet, are of the utmost importance. Insufficient working will leave such structures as those in Figs. 5, 6, and 7, which are detrimental. Proper working produces the good structure of Fig. 8.

**Annealing.** Annealing is carried out at 1550–1675°. In the mill anneal the steel is slowly heated to the desired temperature, held there for an extended period, and slowly cooled. Each of these stages should take something like 8 hr., or at least 24 hr. for the whole cycle, the times and temperatures varying somewhat with the size of the section and the hardness desired for machining. The cooling stage often has to be very much slower and may require 24 hr. The steel must be protected from oxidation as by packing in cast-iron borings or in a mixture of ashes and charcoal. High-speed steel is usually purchased annealed, ready for machining. Reannealing and rehardening of a used tool, or one accidentally tempered too high, is seldom attempted, since



FIG. 5.—Four-in. billet reduced from 7-in. ingot and annealed. Massive carbides in hook form.  $\times 275$ .

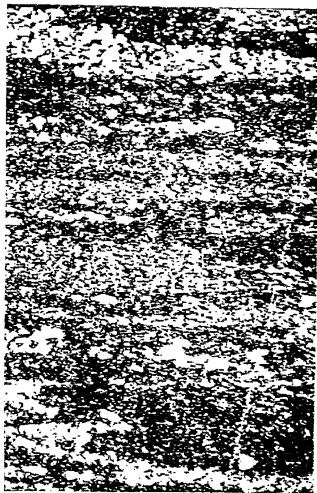


FIG. 6.—Octagon of  $3\frac{1}{2}$  in., hammered from 9-in. ingot. Carbide laminations.  $\times 275$ .



FIG. 7.—Two-in. round from  $7 \times 8$ -in. ingot. Carbide lines still remain.  $\times 275$ .

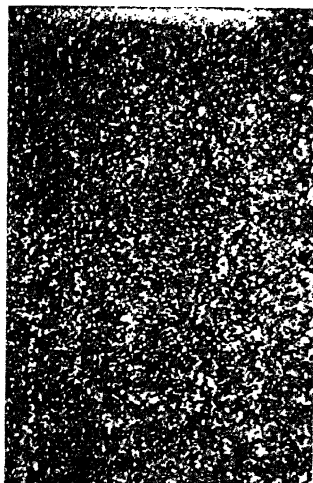


FIG. 8.—Properly annealed steel.  $\times 200$ .

High-speed steel. (MacFarland.)

it is generally thought that a flaky or fish-scale structure is developed which makes the steel brittle.

The hardener, therefore, normally deals with high-speed steels in condition such that no annealing is required prior to hardening. Normalizing is of course avoided since the steels are air hardening.

**Heat Treatment.** Heat treatment consists of preheating the steel slowly up to 1500–1600°, for which a “preheat” furnace or bath is always provided. Then the steel is transferred to the high temperature furnace or bath, which is held at the desired temperature, seldom less than 2200° and seldom more than 2400°, and left there for only a short period. It is desired to get temperature uniformity, but not to give the edges of the tool sufficient time to grow coarse grains and become brittle. To get the proper balance between these two opposing factors, large tools are given longer time in the lower range of temperature, smaller ones a shorter time at higher temperature, though the higher the temperature, short of producing incipient fusion, and without inducing grain growth, the better the tool performance.

When the heating is done in air, as in a muffle, or in the oxidizing products of combustion, the time is limited also by the oxidation of the surface. Small tools at 2350–2400° are heated for say 1½ to 3 min.; at 2300° the time would ordinarily range from 3 to 5 min.

The best results would be obtained at 2400° but that is near the danger line; 2350° is the normally adopted maximum so as to give a bit more leeway in time; and if the furnace is not capable of precise control, 2300° is often adopted for still greater leeway.

**Avoidance of Scaling.** It is desirable to avoid oxidation by using a mildly carburizing atmosphere, or to reduce it by limiting the amount of oxidizing gases in the atmosphere, though the atmospheres are most often kept slightly on the oxidizing side, even when full atmosphere control is used. (Compare Chap. 12, Vol. I.)

Reduction of oxidation is highly desirable, since the burning of the surface in an oxidizing gas raises the surface temperature of the steel and overheats thin edges.

When entire avoidance of scaling is essential and it is desired to raise the C content of the surface and cutting edges, “pack hardening” may be resorted to. The steel is packed in charcoal, or charcoal plus ashes, but without energizer, so that the carburizing action will be mild, heated for ½ to 1½ hr. at 1600–2000°, depending on the desired C pick-up, and its temperature finally raised to about 2000°. This lower temperature and longer time produces much the same core structure, and behavior on hardening, as the shorter times and higher temperatures more commonly used; but it is not a strict alternative be-

cause of the increase in C at the surface, which may make the edges too brittle for some uses.

Salt or lead baths for the high heat would be ideal if they had no decarburizing action and if the melt did not cling to the work to interfere with uniform quenching. The difficulty of avoiding some decarburizing action from these baths at these temperatures, of maintaining the melting pots, and of preventing discomfort or danger from fumes make the method less useful than it would otherwise be.

**Quenching.** The tool, after proper sojourn at the high heat, is quenched in oil, in an air blast, or, for small pieces, merely allowed to cool in still air. Martensite begins to form at 400–350° and continues to form until the temperature has dropped below 200°.

Above 400° the steel may be straightened. It should be cooled to around 200° before tempering so as to secure practically full martensitization.

**Tempering.** The full hardness and toughness are not secured without drawing. The usual draw time is at least 1 hr., preferably 2, and the temperatures range from 1025 to 1100°. The lower temperature gives less secondary hardening and retains some austenite, so that the lower draw leaves the steel softer and tougher than the higher draw, in contradiction to the results in tempering ordinary steels. A long draw, say 8 hr. at around 900°, will give maximum toughness, retaining most of the austenite and tempering the original martensite slightly, but does not develop full hardness.

Tempering above 1100° decomposes the austenite and tempers the martensite, producing progressive softening, but without sufficient increase in toughness to warrant.

Many workers think that repeating the tempering once, or more, using a moderate tempering temperature, say 1025°, the tool being cooled in between, gives better results than a single prolonged draw. Although the evidence is not very clear, the bulk of it indicates that intermittent tempering does give a better and tougher tool. A final draw after grinding the tool to finished dimensions is useful to relieve any stresses thus introduced.

**Composition vs. Heat Treatment.** Heat treatment is altered only to a minor degree when the composition of high-speed steel is changed within the usual limits.

For cutting tools the C is held within narrow limits, around 0.65–0.70%, when the W, Cr, and V approximate the usual 18 : 4 : 1 ratio.

The Cr content seldom deviates much from 4%, lower contents give steels deficient in hardness, higher ones steels deficient in toughness.

The V content has steadily crept up. The earlier steels had 18%

W, 4% Cr, without any V, but in present-day steels 1% V is normally fixed as the minimum, and an increase, to around 2%, is considered to improve cutting ability further, though the first 1% gives more improvement than the second.

**High-vanadium Steels.** When the 18:4:2 ratio is used, the C is generally increased a few points. Still better cutting ability is obtained by raising the C to 1–1.25% and the V to 4%, but of course the cost is materially increased. The W content has been pretty well standardized at 18%. It may be increased to around 25%, probably with slight increase in performance but scarcely in proportion to the added cost.

**Steels with Lower Tungsten.** A decrease of W to 14%, i.e., a 14 : 4 : 1, has been employed, and in roughing cuts such a steel seems to act as well as the 18 : 4 : 1. For fine cuts, however, the diminution in W has to be made up by increase in V, to 14 : 4 : 2, so that the cost saving is small. The 14 : 4 : 1 grade is usually quenched from a lower temperature (around 2300°) than the higher W grade.

**Low Carbon 14 : 4 : 1.** A 14 : 4 : 1 with 1% max. V and with C as low as 0.30% is used, not for cutting tools, but for hot die steels. It would have to be quenched from far too high a temperature, if it were attempted to give it the full hardening necessary for a true tool. It is actually quenched from about 2300° max. and tempered at about 1100°, to 45–46 Rockwell C, for the toughness, combined with red hardness, desired in hot die work.

**Molybdenum in High-speed Steel.** The regular 18 : 4 : 1 may have its W nearly all replaced by Mo. Since Mo is cheaper than W and only about half the percentage is needed, a considerable saving in raw material cost is obtained. The present base price<sup>4</sup> for W high speed is 67 cents per lb., that of equivalent Mo high speed, 54 cents per lb. Since the Mo steel has 8% lower specific gravity, the ratio on a piece basis is 67 to 49.5. Moreover, Mo is found in quantity within the United States borders, whereas nearly all the W is imported. Hence the War Department has been interested in developing Mo high-speed steel. Its efforts, plus those of others, especially Emmons, have overcome the difficulties that were met in the earliest work on such steels.

**Decarburization.** These difficulties are chiefly connected with the more ready decarburization of the Mo steels. The soft skin thus produced was at first thought to be due to loss of Mo, especially as clouds of Mo oxide vapor are given off at forging or hardening temperatures. Actually, however, the loss of Mo is tiny, though that tiny amount makes a lot of smoke. It is loss of C that really counts. By coating the steel with borax during forging and heating for forging, and by

using a suitable controlled atmosphere or other suitable means, such as non-decarburizing fused baths, in heating for hardening, this has been controlled so that only for those tools that have no finish grinding at all need the steel be avoided on this score. Efforts are under way to avoid decarburization by adding other alloying elements to the steel, but it is difficult to accomplish this and still preserve full hardness, cutting power, and toughness. Compare Breeler's <sup>14</sup> data on Cu and B.

**Composition and Heat Treatment of Molybdenum High Speed.** The complete replacement of W by Mo, producing a W-free steel, may be feasible, but more commonly about 1.50% W is retained, the remainder of the W being replaced by 8 to 9% Mo. The Cr and V are left at the usual levels. The C is raised about 0.10% to about 0.75% average. Gill <sup>5</sup> has recently suggested raising the V in a W-free tool steel of 0.81

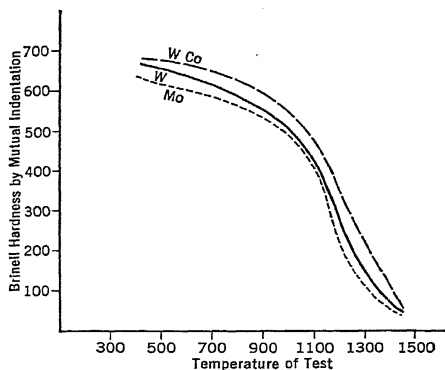


FIG. 9.—Hot hardness of high-speed steels. (Harder and Grove.)

Curve W Co is for a steel of 17W, 4 Cr, 1 V, 4 1/2% Co. Curve W is for ordinary 18W, 4 Cr, 1% V. Curve Mo is for 0.70 C, 9 1/2 Mo, 3 1/2 Cr, 1.20% V.

C, 9.00 Mo, 4.25 Cr, 2.20% V. As usual, the C is raised as the V is increased above the normal.

For the steels with Mo as chief alloying element, quenching temperatures should be slightly reduced; 2350° is the usual top, instead of 2400° as in 18 : 4 : 1 and 2200–2250° is preferred. The lower temperature is, of course, an advantage. The tempering temperature may be slightly reduced also, to about 975–1050°.

The Mo steels have not yet replaced a large tonnage of 18 : 4 : 1 but the replacement is rapidly increasing. The greater care required in heating and the greater difficulty in controlling one more element in an already complex composition have held them back a bit, but experience will make these factors of less moment. For the great bulk of uses the performance seems to be closely of the order of that of

18 : 4 : 1. Some reports show slight inferiority, others slight superiority; the differences are only of the order that would be expected in different lots of 18 : 4 : 1. In tests on dry face cutting of cast iron Kraus and Weddell<sup>6</sup> found that 8% Mo high speed performed slightly better than 18 : 4 : 1, but not quite so well as 18 : 4 : 2.

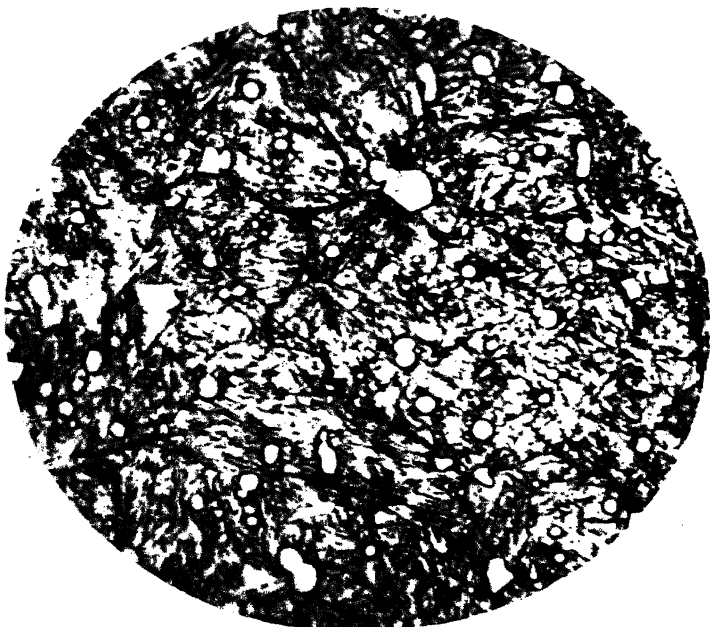


FIG. 10.—Tungsten high-speed steel.  $\times 2000$ . ( $2\frac{1}{2}$  min. etch in 6% Nital.)  
(Courtesy of J. V. Emmons.)

C, 0.70; Mn, 0.34; Si, 0.36; Cr, 3.90; W, 18.00; V, 1.04%. Hardened at 2380° (30 sec. at temperature in salt bath). Tempered at 1040° (30 min. in lead bath). 65.5 Rockwell C.

For very severe service, in which maximum red hardness is required, Mo high speed is generally considered not quite so good as 18 : 4 : 1, since it does not hold its red hardness quite so well, as is shown by the hot hardness data of Harder and Grove,<sup>7</sup> Fig. 9. It should be noted that the hot hardness curve is for a Mo steel without any W at all, instead of the commercial Mo steel which carries a little W. More care is required to keep the Mo steel flooded with coolant. The Mo steel, however, has less propensity toward chipping.<sup>8</sup>

The structure of the Mo steel is indistinguishable from that of 18 : 4 : 1, as Figs. 10 and 11 show. Kennedy<sup>9</sup> gives a series of micrographs showing the structures of the two at comparable variations in hardening and tempering temperatures, which establishes that the two are indistinguishable at all comparable stages. That this is so, in view

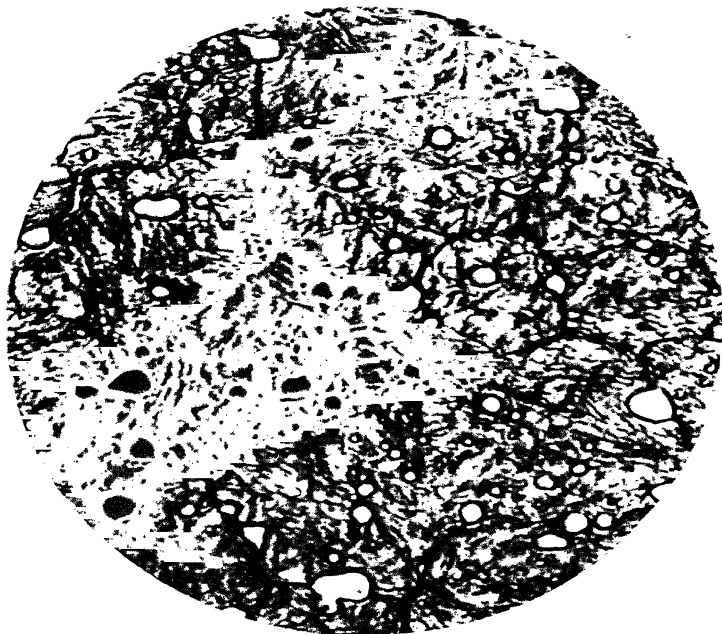


FIG. 11.—Molybdenum high-speed steel.  $\times 2000$ . ( $2\frac{1}{2}$  min. etch in Nital.)  
(Courtesy of J. V. Emmons.)

C, 0.80; Mn, 0.16; Si, 0.29; Cr, 3.65; W, 1.35; Mo, 9.01; V, 1.23%. Hardened at  $2240^{\circ}$  (30 sec. at temperature in salt bath). Tempered at  $1050^{\circ}$  (30 min. in lead bath). 66.5 Rockwell C.

of the higher C content of the Mo steel, indicates that more C is in solution in the matrix, or that it takes more C to produce the carbide.

**Steels with Small Molybdenum Additions.** Small replacements of W by Mo, or the addition of 0.5–1.00% Mo to 18 : 4 : 1, are quite common. The improvement seems to be of the same order as that which would be obtained by increasing the W by a corresponding amount.



At any rate the feasibility of allowing some admixture of Mo in a W steel, or W in a Mo steel, is clear enough so that the reclamation of high-speed scrap does not require the entire segregation of one type of scrap from the other.

Scherer<sup>10</sup> claims that the performance of 18 : 4 : 1 can be practically duplicated or even bettered with the first five steels in the following list:

% C	% W	% Mo	% Cr	% V
0.75	8	1	4	1
0.8*	6	4	4	1
0.8	6	1	4	2.5
0.75	2	8	4	1
1.5	...	7	4	3.5
1.5	...	3	4	4

The last steel is said to be equivalent to 18 : 4 : 0.5.

Houdremont and Schrader<sup>11</sup> corroborate the effectiveness of substituting Mo for W, especially when V is raised to unusually high amounts and C correspondingly raised. Much better cutting performance than that of 18 : 4 : 1 was reported for quite a range of low-W or W-free compositions, among which the following were outstanding:

% C	% W	% Mo	% Cr	% V	Oil Quenched from	Drawn at
1.60	6	4	4	6	2335°	1000°
1.25	6	4	4	4	2335°	930°
1.55	3	3	4	5	2335°	930°
1.60	.....	3	4	2.5	1263°	1000°

In another test a W-free steel of 1.35 C, 3.20 Mo, 4.70 V, 4.30% Cr, oil quenched 2250°, drawn 1000°, showed up better than 18 : 4 : 1 though it was not quite so tough. This V-Cr-Mo steel did not decarburize to a much greater extent than 18 : 4 : 1, and only scaled about half as much when heated in an oxidizing atmosphere for 8 hr. at 2190°.

These German studies were not extensive enough nor did they examine a sufficient range of the different types of machining operations to warrant complete acceptance of the conclusions as to equality of the very low-Mo, W-free steels with 18 : 4 : 1. The evidence is con-

\* Compare discussion in *Steel Horizons* (Allegheny Ludlum Steel Corp.) v. 1, No. 2, 1939, pp. 7-8.

clusive, however, that there is nothing sacred about the 18 : 4 : 1 composition and that, by replacing some, or nearly all, the W by Mo and raising the V and the C, equivalent tools can be produced. Each composition will have its own range of quenching temperature, so heat treatment must be modified to suit the composition.

**Attempts at Elimination of both W and Mo.** Attempts have been made to eliminate the W or Mo entirely and to use only Cr and V, but without success. Recent Russian work with steels of 0.70–1.00 C, 1.5 Si, 9–13 Cr, 1–2.5% V reports<sup>12</sup> a narrow quenching range and one lower than for 18 : 4 : 1, but that even with the best heat treatment the tools were much inferior to 18 : 4 : 1. Although it is not certain that the Cr content might not properly be shifted slightly from 4% in other steels than 18 : 4 : 1, there is practical unanimity that no very great change in Cr content is justified, whatever the shift in the other components.

**Frazer's Appraisal of Alternative Compositions.** Frazer,<sup>13</sup> who considers 18 : 4 : 1 the best all-purpose high speed, appraises the variants of 18 : 4 : 1 for steels of 0.70–0.85% C, and in relation to 18 : 4 : 1, as follows:

% W	% Cr	% V	% Mo	% Co	
18	4	2	0.5	.....	Higher hardenability, more brittle, good for finishing tools and work on abrasive materials.
1.5	3.75	1.1	9	.....	Cheaper, tougher, but not so high in hot hardness.
14	4	2	.....	5	Increasing hot hardness and efficiency over 18 : 4 : 1 as Co and W increase, more brittle. Best performance in heavy cuts.
18	4	2	0.5	7.5	
19	4	2	0.5	12	

Hot die steels of the 1.75 W, 9 Mo, 1% Cr type, lower in C, are also being developed.

**Cobalt in High-speed Steels.** Cobalt improves the cutting efficiency, against very hard materials, of any of the grades of high-speed steel previously mentioned. Up to 10 or 12% may be added, though a smaller addition, around 4%, is more common. As the Co addition increases, the optimum hardening temperature rises till, with 10% added to an 18 : 4 : 1 steel, the minimum recommended is 2375°, and it is better to heat to 2400–2425°. The Co steels require slow pre-

heating; two preheats, one of around 1200° and a second at around 1650°, are often used. Maximum hardness is obtained by drawing at 1050°. A 14 W, 4 Cr, 2 V, 0.50 Mo, 5% Co steel is said to be very satisfactory for interrupted finishing cuts.

The Co steels are extremely prone to decarburization, even in preheating, so precautions to avoid it in heating are essential. These steels are seldom used for tools that are not finally ground to remove the decarburized stock.

**Uranium High-speed Steels.** Uranium is a carbide former with considerable chemical similarity to W and Mo. It has been added to high-speed steels. Such steels were mentioned at some length in the previous edition of this book. These steels have not survived, however, and pending a larger supply and a lower cost of the element, studies have been lacking that would establish whether its use would be worth while if it should become readily available. Such experience as was had with these steels did not bring forth any outstanding properties that cannot be secured with the usual alloying elements that are more readily available.

**Sintered Carbides and Hard Facings.** A discussion of high-speed steels might lead up to consideration of the sintered carbide tools. In these, carbides of W, Ta, Ti, or other hard carbides or mixtures of them, usually bonded with Co, are used to give cutting particles in a tough matrix. In some kinds of service, such tools far out-perform high-speed steels. One might go on from this to consideration of stellite and other alloys containing various proportions of Co, Cr, W, Mo, C, and sometimes boron, used as cast tools, inserted cutting edges, or welded-on overlays for wear resistance, which likewise utilize hard constituents in a more or less tough matrix. Since these are not steels, discussion of them cannot be included here. Selected references to these topics will suffice.<sup>13</sup>

## BIBLIOGRAPHY

1. GROSSMANN, M. A., and E. C. BAIN: "High Speed Steel." New York, J. Wiley and Sons, Inc., 1931, 178 pages.
2. GREGG, J. L.: "The Alloys of Iron and Tungsten." New York, McGraw-Hill Book Co., 1934, 511 pages.
3. ANDREW, J. H., and G. W. GREEN: The Manufacture and Working of High-speed Steel, *J. Iron Steel Inst.*, v. 99, 1919, pp. 305-344.
4. *Daily Metal Trade*, September 23, 1938.
5. GILL, J. P.: Alloy Tool Steel U.S. Patent 2,105,114, January 11, 1938.
6. KRACS, C. E., and R. R. WEDDELL: Determining the Tool-life Cutting-speed Relationship by Facing Cuts, *Trans. Am. Soc. Mech. Eng.*, v. 59, 1937, pp. 555-558.

7. HARDER, O. E., and H. A. GROVE: Hot-hardness of High-speed Steels and Related Alloys, *Am. Inst. Min. Met. Eng.*, Tech. Publ. 497, 1933, 39 pages.
8. GUNETTI, H. C.: Some Interesting Problems in Machine Cast and Forged Steel, *The Review* (Am. Soc. Metals), v. 11, January, 1938, p. 2.
9. KENNEDY, R. G., JR.: The Micro-structures of Mo-Max and 18-4-1 High Speed Steels Compared, *Metals & Alloys*, v. 8, 1937, pp. 289-293.
10. SCHERER, R.: Versuch mit wolframarmen und wolframfreien Schnellarbeitsstählen (Studies with Tungsten Poor and Tungsten Free High Speed Steels), *Stahl u. Eisen*, v. 57, 1937, pp. 1355-1359.
11. HOUDREMONT, E., and H. SCHRADER: Neuere Entwicklungen auf dem Gebiete spurstoffarmer Schnellarbeitsstähle (New Developments on the Role of the Scarcer Materials for High Speed Steels), *Ibid.*, v. 57, 1937, pp. 1317-1322.
12. VLADISLAVOV, V., and A. IVANOV: *Kach Stal*, v. 5, No. 4, 1937, pp. 7-11.  
JHITNIKOV, D.: *Ibid.*, v. 5, No. 12, 1937, pp. 24-29.
13. FRAZER, W. R.: Modern Tools of High Speed, *Metal Progress*, v. 29, December, 1936, pp. 41-45.  
COMSTOCK, G. J.: Cemented Tungsten Carbide, *Trans. Am. Soc. Steel Treat.*, v. 18, 1930, pp. 993-1008.
14. BREELER, W. R.: Development in Molybdenum High Speed Cutting Steels. Preprint No. 11, Am. Soc. Metals, October, 1938, 20 pages.
15. McKENNA, P. M.: Tantalum Carbide Tool Compositions, *Am. Inst. Min. Met. Eng.*, Tech. Publ. 897, Metals Technology, v. 5, 1938, 12 pages.  
AGTE, C., and K. BECKER: "Hartmetallwerkzeuge," Berlin, Verlag Chemie, 1937, 234 pages.

## CHAPTER 24

### PERMANENT MAGNET STEELS AND THEIR HEAT TREATMENT

**Soft and Hard Magnetic Materials.** There are two types of steels whose magnetic properties are utilized. To the first belong the soft, easily magnetized steels for such uses as transformer cores, in which easy magnetization and low energy loss of the magnetizing current are desired, together with ability to be rolled into sheet. Especially purified iron shows easy magnetization—i.e., high permeability, but its electric resistance is so low that large eddy currents are induced in it during magnetization. For a balance, very low C, and high Si (up to around 4% Si) to give higher resistivity, are used, and it is found that large crystal size, and a particular orientation of the grains are favorable factors.<sup>1,2</sup> Full annealing in a non-oxidizing atmosphere and process annealing of deformed material for stress relief are the only heat treatment applied to these magnetically soft materials. Special alloys, usually very high in Ni, have specific magnetic properties of particular interest, but, since these are not steels, they will not be discussed here.

The "soft" magnetic materials have been discussed by Schumacher and Ellis<sup>3</sup> and treatment in a magnetic field by Bozorth and Dillinger.<sup>4</sup>

The second type, the members of which are affected by heat treatment comprises the magnetically hard materials used for permanent magnets. Permanent magnets provide a source of magnetism without the use of an external magnetic field and are used in magnetos, relays, magnetic chucks, electric meters, etc.

High-C steel in the martensitic state retains some magnetism after being magnetized, but relatively little and this little is not well retained, so alloying is resorted to.

**Magnetic Testing.** The hardened steel is made into a permanent magnet by putting it in a magnetic field, as by winding it with turns of Cu wire and passing current through the coil. After the current is cut off the steel remains magnetized, and high intensity of this residual or remanent induction is one of the sought-for properties. The magnet does not lose its magnetism as fast when the magnetizing field is re-

duced to zero, as it gains it while the magnetic field is being applied. The plotted curve thus shows hysteresis, and the hysteresis loop obtained in testing evaluates the magnetic properties, much as a stress-strain diagram evaluates mechanical properties. If a negative magnetic field is applied to the magnetized specimen, it can be demagnetized and the coercive force that has to be applied to do this is a measure of the resistance to demagnetization.

The remanent induction is measured in gaussses and, in the terminology of magnetic testing is denoted by Br, while the coercive force is measured in oersteds and denoted by Hc.

In well-treated W magnet steel, Br is over 10,000 and Hc around 65. The product,  $Br \times Hc$ , is often used to express the value of a magnet steel in a single figure. The magnetizing force, normally around 300 oersteds, used in determining these values, affects the values somewhat, and the special magnet steels of high coercive force do not show their true superiority unless high magnetizing force, say 1000 oersteds, is used.

Not only must the initial  $Br \times Hc$  be high, but this product should remain constant on aging. Br varies in commercial magnets only from around 6500 to 12,000; Hc varies from 40 to over 900.

The details of magnetic testing are too specialized for discussion here, and values for Br and Hc are given only as an index to the magnetic quality of an alloy or the suitability of its heat treatment. Testing methods for magnet steels are laid down by the American Society for Testing Materials (A-34-33), and are described by Sanford<sup>5</sup> and by Neumann.<sup>6</sup>

**Permanence.** Since the magnetic condition is a strained one, and strains tend to relieve themselves in time, magnets may lose their strength. Since in a calibrated instrument this is to be avoided, choice of alloy and of heat treatment to produce stability is important, and some of the initial magnetism may be intentionally removed in order to get the magnet down to a reasonably stable level.

**The Size Problem.** The magnetic properties desired in a magnet steel are not quite the same for magnets of different sizes and shapes, that is, the behavior of the magnet depends on its geometry as well as on the properties of its material. The advantages of the "extra good" magnet steels are quite largely in the fact that they have high magnetic strength so that a small magnet will exert as strong a field as a larger one of lower strength, a great advantage in aircraft instruments and the like, but of no great moment if space and weight are not objectionable.<sup>7</sup>

**Stressed Ferrite.** The permanent magnets are extremely hard, fine grained, and contain C. cementite, or other hardening particles in an extremely fine subdivision.<sup>3</sup>

The basis of the steel must be ferrite, in order to be magnetic, but this ferrite apparently needs to be in a peculiar, very localized, state of stress, probably associated with the atomic forces and lattice distortion existing at and near the contact surfaces between like and differing crystals.

Such stresses exist in martensite, and the martensite is usually produced by water quenching, this severe cooling, as has been discussed in Chap. 6, Vol. I, being prone to set up stresses due to dimensional changes in the transition from austenite to martensite.

**Heat Treatment.** It is generally considered that the minimum of heat treatment necessary to harden the steel produces the best magnet that steel is capable of making.

Homogenization treatments, annealing, and the like prior to hardening that set up conditions minimizing the production of stress are avoided in heat treatment of permanent magnets. Similarly, stress relief by tempering of the martensite is held to a low level, since it happens that "magnetic hardness," which is desired, goes pretty much hand in hand with indentation hardness, though the causes may not be identical.

**Austenite.** It seems to be necessary for best magnetic properties to so alloy and quench a magnet steel that some residual austenite is produced, and left untransformed after the quench. Obviously austenite itself, being non-magnetic, would not be useful in a magnet, and its importance must be in the state of stress that accompanies its presence, or in the boundary forces exerted at the junction of austenite and martensite grains. Many of the observations on, and methods for, obtaining magnetic quality arise from cut-and-try experiments and are difficult to connect up with the underlying cause.

If the presence of austenite is necessary, obviously tempering that would convert austenite to martensite or troostite will be harmful. As a matter of fact, magnets are tempered only at boiling water temperature, not higher. Carbon and alloy contents and quenching media are chosen so that the proper amount of, but not too much, austenite is retained, and the tempering or aging process is controlled to prevent its breakdown.

**Martensite.** Quenched martensite has a tetragonal crystal structure (alpha martensite) but on long standing at room temperature, or on tempering at boiling water temperature, it changes to the cubic

structure (beta martensite). This change, probably allowing some stress relief, is accompanied by a small loss of magnetic properties; but once it has occurred, the properties are quite stable. When stability is wanted the magnets are boiled. It is not definitely known whether C, in forced solid solution in alpha martensite tempered only at 212°, exists as finely divided free-C particles, or as finely divided cementite. At higher tempering temperatures cementite, if not already existent, forms and then agglomerates, in plain C steels. The conditions in steels containing other carbide-forming elements beside Fe are even less certain. Austenite, on long tempering at about 265°, breaks down somewhat. To prevent this magnets are tempered at a lower temperature and are not used at elevated temperatures.

Annealing for machinability is carried only to the extent absolutely necessary; full softening is not desired. Re-heat-treatment of a magnet is likely to give properties poorer than the original ones. Some authorities advocate forming the magnet on the original rolling or forging heat so as to avoid reheating, and cast magnets are liked because they do not have to be reheated for forming.

Slow cooling after annealing is detrimental. Long holding at hardening temperature is to be avoided; ten minutes at temperature is considered the optimum by Schulz and Junge.<sup>8</sup>

**Alloying Elements.** The alloying and heat treatment of permanent magnet steels are designed to produce a hardened ferrite, with stress-producing particles dispersed in it. The ferrite-strengthening elements Co and Ni may enter the matrix, and C, carbides, or metallic compounds serve to produce the particles.

The C or cementite in the martensite is made of the type and dispersion desired by introduction of carbide formers, of which W is the best known for the purpose. Because of its similarity to W, Mo might be expected to serve as well. Gregg<sup>9</sup> points out that the cheaper Mo-steel magnets are quite satisfactory. For still cheaper magnets Cr steel is largely used. Besides W, Mo, and Cr, or combinations thereof, and high C, 0.60–1.00%, to supply C for the carbides, the ferrite is advantageously alloyed with quite large amounts of Co, and Ni sometimes appears, though it is not common in magnet steels for martensitic hardening.

Since Co is expensive, the newer high-Co magnets are not so cheap on a performance basis as the older W or Cr-steel magnets. Plain C steels, with the Mn slightly raised, find some use, but the economic balance usually calls for at least the use of Cr, if not W, as an alloying element.



Cobalt and Ni, in large quantities, also appear in the matrix of precipitation-hardened magnets, in which C is kept as low as possible and other compound-forming elements used instead. These high alloys are not steels in the true sense of the word. We shall recur to them after discussing the martensitic magnet steels.

To get hardenability and a little austenite retention, plain C magnet steels usually contain about 0.60% C and 0.80% Mn and are water quenched from about 1450°. The properties thus attainable are only of the order of 9000 Br, 40 Hc. In other magnet steels the Mn is kept low.

**Chromium Magnet Steel.** With C at about 0.60-1.00% and the use of around 1 to 4% Cr, a water or oil-hardening magnet steel is produced of around 9000-9500 Br with Hc ranging from 58 to 70. Both Br and Hc rise with the Cr content. About 1470° (water) for the low Cr and 1525° (oil) for the high Cr are accepted quenches, though the 1% Cr steel, according to some authorities, is better when oil quenched.

The senior author (D.K.B.) has produced many Cr magnets of 0.95-1.00% C, and 3.40-3.60% Cr, which give the best Hc  $\times$  Br of the Cr series, only about 10% less than the usual W magnet steel. By increasing the size of the magnet 10%, the far cheaper Cr steel could be substituted. In the shop the Cr steel was found to be less prone to crack on hardening and gave more uniform results than W steel. The Cr steel supplied by different steel mills was quite variable in behavior, and the steel maker has to take great care in casting, in working, and in annealing the steel to soften it for shearing if uniform magnetic properties are to be obtained.

In spite of the usual belief that each annealing considerably lowers the magnetic properties, it was found that, after the magnets were formed, they could be heated to 1420°, just under  $A_{c1}$ , for the 1 C, 3.50% Cr steel used, and air cooled with no appreciable depreciation of the magnetic properties from those of the steel properly mill-annealed but without subsequent annealing. Around 1% W or 0.35% Mo is sometimes added to the Cr steels.

**Tungsten Magnet Steels.** When the space into which the magnet is to be fitted is restricted so that a still more expensive steel can be paid for, a W steel may be used. Experience has taught that the composition should be in the ranges 0.50-0.80% C usually 0.65-0.75%, and 5 to 7% W, and that the steel should be water quenched.

A little Cr, 0.35 to 2%, may be added, when oil hardening becomes

feasible. The magnetic properties are not notably changed by these small additions.

**Heat Treatment of Tungsten Magnet Steels.** The 6 W, 0.70% C type, with or without Cr, should not be overheated or held too long at the highest temperature in the suitable range, which is 1475–1650°. If “T. Max.” (see Chap. 6, Vol. I) is reached, the austenite is made too sluggish and the steel is not in the right condition for hardening for magnetic properties. It is possible that some very tiny undissolved cementite particles should remain in the austenite before quenching.

In W steels of this composition, annealing for maximum machinability, or hot rolling at too low a finishing temperature, separates out, as the carbide WC, the W contained in cementite. This carbide does not readily go into solution on heating for hardening, and a steel thus annealed before hardening has poor Hc. The separation of WC and refusal to harden properly are termed “spoiling.” Spoiled steel may be reclaimed by normalizing from 2100° or by rerolling to a smaller size. The hot work helps to take the WC into solution, but such reheating is avoided wherever possible.

The presence of Cr tends to prevent “spoilage,” and where magnet steels must be reheated in fabrication the Cr addition is especially desirable.

The steels with W, or W with small Cr addition, give 10,000–11,000 Br and 60–70 Hc. When properly handled, as high as 80 Hc has been reported for 6 W, 2% Cr. For permanence, tempering at 212° is resorted to. (Usually W magnets are alternately boiled and then cooled in ice water, the cycle being several times repeated instead of continuous 212° tempering.) This slightly reduces the properties but makes them more stable. Part of the deterioration of a magnet is ascribed to mere jarring or to temperature oscillations, and this treatment is designed to produce stability against that type of disturbance as well.

Quenching while the steel is in a strong magnetic \* field so that the martensite is magnetized the instant it forms is said to produce markedly beneficial results, but this is not general practice. Dowdell<sup>10</sup> advocates hardening, magnetizing, demagnetizing, boiling for 12 hr., remagnetizing, boiling again till 5% of Br is lost, then reversing the magnetic field until another 5% has been lost.

**Commercial Magnet Steels.** The requirements for magnets and the properties of various magnet steels have been discussed by Scott,<sup>11</sup> who gives the following table for the steels hardened by martensitic quenching.

## COMMERCIAL TYPES OF MAGNET STEELS (AMERICAN)

Type of Steel	NOMINAL COMPOSITION, % <sup>a</sup>					MAGNETIC PROPERTIES <sup>b</sup>		Hardening Temp.	Quenching Medium
	Mn	Cr	Co	Mo		Br Gauss	Hc Oersted		
C-Mn	0.60	0.80				10,000	43	1450	Water
Cr	0.60	0.40				10,000	50	1450	Oil
	0.90	0.35				9,000	58	1500	Water, Oil
	0.95	0.20 to 0.60				9,500	63	1525	Oil
	1.00	0.35				9,000	72	1550	Oil
Cr-Mo	1.00	0.35			0.35	9,000	65	1550	Oil
W	0.70	0.30	0.20	5.5		10,000	65	1550	Water
W-Cr	0.70	0.50	0.50			9,000	70	1525	Oil
Co	0.90	0.35	4.75	1.25		7,500	120	1650	Oil
	0.90	0.30 to 0.85	3.50 to 5.75	3.75 to 7.0	0.35 to 41	9,700	235	1700	Oil
Co-Cr	0.95	0.30	9		16	8,000	180		Air
Co-W	0.85	0.50	2 to 5	8.75	17	9,000	165	1750	Oil

<sup>a</sup> Silicon limit usually 0.30% max., P and S 0.03% max., Ni 0.50% max.

<sup>b</sup> Nominal values after hardening. Values for previously annealed materials usually 5 to 10% lower in Hc.

<sup>c</sup> Triple heat treatment, 2100° F. cool in air 1150° F. cool in air 1850° F. cool in air.

Dowdell<sup>10</sup> recommended three magnet steels chosen from a large number on the market at the time of his investigation. These were:

	% C	% Mn	% Si	% P	% S	% Cr	% W	% Mo	% Co
1.....	0.86	0.13	0.16	0.015	0.041	9.31	.....	1.24	13.66
2.....	0.97	0.74	0.33	0.031	0.005	2.87	.....	.....	.....
3.....	0.76	0.61	0.22	0.022	0.011	0.41	5.08	.....	.....

Of these, No. 2 is by far the cheapest. According to the experience above mentioned, which is consistent with the data of Gumlich,<sup>12</sup> though his tests did not cover the gap between 3% and 6% Cr, the most economical composition in the Cr magnet steels for high Hc × Br is at about 1 C. 3.5% Cr. According to Kinzel and Crafts,<sup>13</sup> the Cr magnet steel most used at present runs 1% C, and 2 to 4% Cr, as giving the best economic balance.

**High Cobalt Additions.** Addition of 8–15% Co to the Cr-W or Cr-Mo magnet steels raises the quenching temperature to around 1650° (oil), and while the Br may drop somewhat, say to 7500, Hc increases to around 120–190.

Honda and co-workers hit upon a new group of magnet steels in which there is even a larger jump in Hc through larger replacement of

Fe by Co. With the C about the same as in the other magnet steels, and the W and Cr at the high end of their respective ranges for the simple steels, the usual composition becomes 0.75–0.90 C, 7–9 W, 2–5 Cr, 35–41% Co, though part or all of the W can be replaced by Mo, the Cr can be raised and the W lowered, or the Mn may be raised. Such steels are oil quenched from about 1700–1750°. Br remains at around 9000–10,500, but a very large increase is made in Hc, from 60, 70, or at most 80, in the Fe-W-Cr steels without Co, to 220–270. Since the merit index is  $Br \times Hc$  this means that these "KS" steels, as Honda christened them, are 3 times as good. Similar steels with 15–17% instead of 35% Co, are in large use. The composition is generally 17% Co with 8 to 9% W.

**Precipitation Hardenable Alloys.** A still further increase in Hc over that of the 35% Co martensitic steels is attainable in a group of alloys that depend on precipitation of a metallic compound for their hardening. The development of these is largely due to Mishima,<sup>14</sup> Köster,<sup>15</sup> and Seljesater and Rogers.<sup>16</sup>

These steels are preferably very low in C. As a base, all these alloys contain from 18 to 30% Ni and from 50 to 69% Fe. The remainder may be W, Al, V, Co, Mn, Cr, Cu, and Mo in varying combinations and percentages. Two popular groups include the "Alnico" type consisting of 16–30 Ni, 7–12 Al, 5–10 Co, 52–69% Fe, and the "Alnic" group containing 18–30 Ni and 10–12% Al, the remainder being Fe.

Alloys of this type may be heat treated in one of two ways. Both methods involve a solution treatment at a very high temperature, of the order of high-speed steel-hardening temperatures, say 2300°. In one method this solution treatment is followed by rapid cooling and a precipitation treatment at 1200–1300°. The other method, which has been developed principally for the Alnic group, involves cooling from the solution temperature at an experimentally determined rate.

In both methods precipitation is carried further than will give the highest indentation hardness, i.e., to a point where the separated compound is agglomerated sufficiently to be visible under the microscope. The precipitation hardened material is alpha iron plus Fe-W-Co, Fe-Mo-Co, Fe-Ni-Al, or Ni-Al-Co compounds. Of the various precipitation-hardening alloys the Alnic (Fe-Ni-Al) alloys have received the most attention. After a detailed X-ray analysis of these alloys, Bradley and Taylor<sup>17</sup> confirmed the idea that the composition corresponding to the formula  $Fe_2\text{-Ni-Al}$  is the most desirable one in this system. They advanced the hypothesis based on experimental evidence that this composition may exist in two forms. Both forms are magnetic and

have body-centered cubic lattices. However, one form, which is obtained by rapid cooling, has the atoms of the various elements arranged at random in the lattice. In the other form, obtained by slow cooling, the atoms at the cube corners are different from those which occur at the cube centers. This type of lattice is known as a "super lattice," and in this case the lattice spacing is 0.3% greater than in the normal lattice. When a proper cooling rate has been found, or when the heat treatment consists of rapid cooling from a solution temperature followed by a precipitation treatment, both lattices exist simultaneously and the difference in volume between the two is responsible for the internal stresses which produce high values of  $H_c$ . In other alloys of the precipitation-hardening type, the mechanism producing high values of  $H_c$  may be similar to the one observed for the Fe-Ni-Al alloys, although details on other alloys have not yet been worked out.

Since the compounds, or differences in lattice structure, responsible for the magnetic properties of these alloys are formed at a high temperature, magnets made from these alloys have a high degree of permanence and may be used at much higher temperatures than the martensitic types. A coercive force of 800–900 oersteds has been reported for such alloys, though with somewhat lower values of remanence than in martensitic steels. However, with remanence at the same level as the martensitics, 400–500  $H_c$  can be reached.

The "merit index" thus rises from  $9000 \text{ Br} \times 40 \text{ Hc} = 360,000$  in the C steel magnets to say  $7500 \times 800 = 6,000,000$  in some of the precipitation-hardening types. Thus magnets of the Alnico and Alnic types occupy much less space for a given strength than the martensitic types.

The precipitation hardening alloys are relatively new, and the most desirable compositions and heat treatments are not yet thoroughly worked out.

#### BIBLIOGRAPHY

1. BITTER, F. C.: "Introduction to Ferromagnetism." New York, McGraw-Hill Book Co., 1937, 314 pages.
2. GREINER, E. S., J. S. MARSH, and B. STOUGHTON: "The Alloys of Iron and Silicon." New York, McGraw-Hill Book Co., 1933, 457 pages.
3. ELLIS, W. C., and E. E. SCHUMACHER: Magnetic Materials, *Metals & Alloys*, v. 5, 1934, pp. 269–276; v. 6, 1935, pp. 27–29.
4. BOZORTH, R. M., and J. F. DILLINGER: Heat Treatment in a Magnetic Field, *Metal Progress*, v. 30, September, 1936, pp. 82–86.
5. SANFORD, R. L.: Some Principles Governing the Choice and Utilization of Permanent-Magnet Steels, U.S. Bur. Stand. Sci. Paper 567, 1927, 11 pages. See also: Bur. Stand. Circular 415, Magnetic Testing, 1937.

6. NEUMANN, H.: Magnetische Messverfahren und Prüfgeräte für Dauermagnetwerkstoffe (Magnetic Measurements and Testing Methods for Permanent Magnet Materials), *Archiv f. Eisenhüttenw.*, v. 11, 1938, pp. 483-496.
7. Fast Heating in Shell Production, *Steel*, v. 101, December 13, 1937, p. 29.
8. SCHULZ, E. H., and W. JUNGE: Beitrag zur Frage der Wärmebehandlung und Prüfung von Chrom-Magnetstahl (Heat Treatment and Testing of Chrome-Magnet-Steel), *Stahl u. Eisen*, v. 46, 1926, pp. 11-13.
9. GREGG, J. L.: "The Alloys of Iron and Molybdenum." New York, McGraw-Hill Book Co., 1932, 507 pages.
10. DOWDELL, R. L.: Investigation of the Treatment of Steel for Permanent Magnets, *Trans. Am. Soc. Steel Treat.*, v. 5, 1924, pp. 27-66.
11. SCOTT, K. L.: Steels and Alloys for Permanent Magnets, *Metal Progress*, v. 30, September, 1936, pp. 64-68, 88.  
SCOTT, K. L.: Magnet Steels and Permanent Magnets, *Trans. Am. Inst. Elect. Eng.*, v. 51, 1932, pp. 410-417.
12. GÜMLICH, E.: Untersuchungen an Chromkohlenstoffstählen für permanente Magnete (Studies on Chromium-Carbon Steels for Permanent Magnets), *Stahl u. Eisen*, v. 42, 1922, pp. 41-46.
13. KINZEL, A. B., and W. CRAFTS: "The Alloys of Iron and Chromium." New York, McGraw-Hill Book Co., v. 1, 1937, 535 pages.
14. MISHIMA, T.: Magnet Steels Containing Nickel and Aluminum. Strong Permanent Magnet with Tungsten, Molybdenum, or Chromium. Strong Permanent Magnet with Cobalt. Permanent Magnet Containing Copper. Permanent Magnet Containing Nickel, Aluminium, Cobalt, and Chromium. Permanent Magnet Containing Nickel, Aluminum and Manganese. Permanent Magnet Containing Nickel, Aluminum, and Vanadium. U.S. Patents 2,027,994-2,028,000, January 14, 1936.
15. KÖSTER, W.: Das System Eisen-Nickel-Aluminium (The System Iron-Nickel-Aluminium), *Archiv f. Eisenhüttenw.*, v. 7, 1933, pp. 257-262.
16. SELJESATER, K. S., and B. A. ROGERS: Magnetic and Mechanical Hardness of Dispersion Hardened Iron Alloys, *Trans. Am. Soc. Steel Treat.*, v. 19, 1932, pp. 551-576.
17. BRADLEY, A. J., and A. TAYLOR: *Nature*, v. 140, 1932, p. 1012.



# INDEX

NOTE: Numbers in light-face type refer to Vol. I; numbers in bold-face, to Vol. II.

- Abnormal steel in carburizing, 219, 266, 281
- Ac, S1
- Ac<sub>1</sub>, importance of exceeding, for grain refinement, 96
- Activated baths for carburizing, 349
- Age hardening, 203
- Aging, **14**
  - cast steel, 203
  - quench, 206
  - strain, 207
- Air cooling (normalizing), 105
- Air-gas ratio in preparing controlled atmospheres, 325
- Alloy "extras" cost of alloy steels, **322**
  - hot-rolled, cost, **322**
- Alloy steels, automotive uses, choice of, **332**
  - definition of, 2; **322**
  - heat-treating methods, relation to choice of, **324**
  - properties, comparison of, **327**
  - replacement of one by another, **326**
- Alloying elements, cost, **213**
  - effect in steel, **177**
  - effect on eutectoid composition, **194**
- Alloys, effect of, on critical cooling rates, 161
- Alpha iron (ferrite), 86
  - body-centered structure, 86
- Alumina, in steel, effect in carburizing, 266
- Aluminum, additions for grain-size control, 220, 236
  - effect in free-cutting steel, **54**
  - free nitriding steels, **295**
  - in Cr-Mo steel, **367**
- Aluminum, in medium Mn steel, for low temperature use, **383**
  - in Mn pearlitic steel castings, **80**
  - in Mn steel, **73**
  - in nitriding steels, **295**
  - in steels for grain-size control, **157**
- Aluminum-chromium-molybdenum steel for nitriding, **298**
- Ammonia, cracked, as protective atmosphere, 299
  - dissociation of, in nitriding, 254, 257, 362
  - for nitriding steel, 351
- "Amola" steel, **172, 173**
- Annealed steel *vs.* quenched and toughened, 198
- Annealing, 94
  - atmospheres for, 298
  - box, *see* Box annealing.
  - bright, *see* Bright annealing.
  - C tool steel, 124
  - cold-worked steel, 93
  - cooling from, 102
  - cooling rate after, 115
  - covers, 402
  - decarburization in, 120
  - double treatment, 112
  - effect of, on structure, 95
  - for maximum ductility, **114**
  - furnace cooling in, 104
  - heat application, 94
  - medium-C steel, 98
  - of packs and coils, 416
  - process, grain growth in, **14**
  - relation of hardening to, 150
  - semi-hard (0.55-0.75% C) steel, 120
  - steel castings, 100
  - strip and sheet, 134
  - summary of methods, 122



- Annealing, temperature, 89, 100  
     *vs.* normalizing, 141  
     *vs.* toughening, 198  
 Ar<sub>1</sub>, depression of, 89  
 Ar<sub>2</sub>, 89  
 Ar' (Ar prime), 155, 156  
 Ar'' (Ar second), 156  
 Armor plate, carburizing of, 275  
 Arrests in heating and cooling curves,  
     81, 88  
 Atmospheres, circulation of, in fur-  
     naces, 415  
     controlled, *see* Controlled atmos-  
         pheres.  
     for annealing, 298  
     for heating high-speed steel, 293  
     for normalizing, 301  
 Austempering, carburized steels, 288  
     definition of, 143  
     spring steels, 310  
     temperature for, 144  
 Austenite, cementite absorption by, on  
     heating, 88  
     coarse-grained, sluggishness, 237  
     coarsening of, on heating, 223, 227  
     coarsening temperature of, 231  
     definition of, 82  
     ferrite absorption by, on heating, 84  
     residual, in magnet steels, 462  
     retained, effect on dimensional  
         changes, 179  
     retained in carburized steel, 285;  
         271  
     retained in Mn steel, 73  
     retained in Ni steel, 121  
     sluggishness produced by alloys, 180  
     solubility of C in, 82  
 Austenitic steels, 396  
     carbide separation in, 397  
     compositions of, 396, 399  
     for high temperature service, 371  
         Al and Ti in, 373  
         Mo and W in, 371  
     Mn, 399  
     Mn, rails of, 400  
     Ni, 398  
     properties, 396  
     stabilization by Ti, Cb, etc., 398  
 Autofrettage, 183  
     Mo-V and Ni-Cr-Mo steels for, 201  
 Automatic control, 432  
 Automatic furnaces, 431
- ### B
- Banded structures, elimination of, 113  
 Batch furnaces, 410, 424  
 Baths, *see* Individual types, *i.e.*, Salt  
     baths.  
 Battering tools, 436  
 Bearings, Cr steel, 141  
 Bend tests, 29, 55  
     on austempered stock, 144  
 Bessemer steel, P and N in, 52  
 Beta iron, 86  
 Blueing, 298  
 Bolting stock, 376  
     load-carrying ability, 376  
     relaxation properties, 377  
 Borings, cast-iron, packing in, for pro-  
     tection of surface, 332  
 Boron in steel, 161  
 Box annealing, 121  
     atmospheres for, 298  
 Bright annealing, 301  
 Brinell hardness testing, 8  
     table of impressions, diameters *vs.*  
         Brinell numbers, 10  
 Brinell strength *vs.* endurance limit, 37  
 Brinell *vs.* tensile, Fig. 8, 21  
 Brittleness, from grain growth after  
     critical strain, 121  
     in impact tests, 44  
     Stead's, 121; 14  
 Burning of steel in forging, 386  
 Butane for gas carburizing, 273
- Calcium cyanide in salt baths for cya-  
     niding, 349  
 Calorizing, 372  
 Carbide, balance, 338  
     formers, Cr steel, 251  
     forming elements, effects, 179, 181  
     precipitation in Cr steel, 208  
 Carbides in C tool steels, 421  
     in intermediate tool steels, 433  
     Mo and W, stability, 170  
     sintered for tools, 458

- Carbon, content of carburized case, factors affecting, 261
- content of carburizing steels, 252
- diffusion of, in cyaniding, 349
- effect on normalized and annealed steel, 4
- effect on strength, 81
- gradient in carburized case, 259
- range, 5
- solubility in austenite, 82, 252
- tool steel, *see* Tool steels, carbon.
- vs.* alloy content of carburized steel, 282
- vs.* toughness, 198
- Carbon dioxide, effect in controlled atmospheres, 328, 331
- in gas carburizing, 263
- removal of, 328
- scaling by, 310
- Carbon-manganese steel, springs, 32
- Carbon monoxide, in gas carburizing, 263
- inactivity, 308
- production, 321
- Carbon monoxide-carbon dioxide, ratio in controlled atmospheres, 307
- Carbon steels, 1
- annealing and normalizing, 8
- average properties, 39
- castings *vs.* forgings, 7
- eutectoid, 32
- grain growth in, 32
- grain size, 10
- high, inclusions in, 33
- notch sensitivity, 33
- limitations, 9
- machinability, 13
- oil *vs.* water quenching, 43
- quenching temperature, 11
- size effect, 10
- small sections, oil quenched, 11
- springs, 32, 302
- vs.* Mo steel at high temperatures, 361
- Carbon steels under 0.15%, 12
- quenching, 13
- Carbon steels 0.15–0.25%, 15
- annealing, 17
- effect of Mn in, 16
- quenching, 15
- Carbon steels 0.20% C, limitations, 209
- Carbon steels 0.25–0.35% C, 19
- annealing, 22
- heavy forgings of, 21
- quenching temperature, 21
- Carbon steels 0.35–0.45% C, 27
- heavy forgings of, 28
- quenching temperature, 30
- Carbon steels 0.45–0.60% C, 30
- annealing, 31
- quenching and tempering, 30
- Carbon steels 1% C, 33
- Carburization, mechanism of, 261
- Carburized case depth *vs.* alloy content of steel, 282
- Carburized steels, austenite in, 271
- case and core properties *vs.* heat treatment, 281
- case and core testing, 288
- case and core toughness, 285
- core properties, 273
- distortion, 291
- double quench, 282
- endurance, 290
- grain growth in, 279
- hardness of, 284
- heat treatment, 280
- impact on, 285
- properties, 288; 285
- single quench, 282
- tempering, 284; 274
- Carburizers, 266
- leather as, 267
- liquid, 348
- regeneration of, 267
- Carburizing, 251, 353
- abnormal steel in, 219, 266, 281
- armor plate, 275
- box, oxygen absorption in, 266
- pack, 253, 264
- boxes, 268
- packing of, 269
- compounds, 263, 266
- decarburization in, process, 264
- diffusion in, 257
- effect of alloys on, 255
- effect of Al or  $Al_2O_3$  in steel for, 266
- energizers for, 264
- furnaces for, 271

- Carburizing, gas, 271  
     packing charge for, 416  
     grain size in, 253, 281  
     hydrocarbons for gas, 274  
     dilution with fuel gas, 273  
     mixed gases, 274  
     nitriding processes, 353  
     normal steel in, 219  
     prevention of, 255  
     rate of, *vs.* time and temperature, 255  
     S.A.E. steels, 285  
     steel, C content, 252  
     temperature control in, 270  
     tendencies of controlled atmospheres, 328  
     use of charcoal in, 265
- Carburizing steels, case properties, 271
- Cr, 137  
     core properties and composition of steel, 278  
     high-C for thin cases, 273  
     Ni in, 122  
     machinability of, 254  
     S.A.E. steels, 274  
     0.30% C, 274
- Case carburizing, *see* Carburizing.
- Case depth, measurement of, 258  
     time and temperature, effect of, in cyanide and activated baths, 349
- Case properties of carburized steel, *see* Carburized steel.
- Case testing of carburized articles, 289
- Cast-iron borings, *see* Borings, cast-iron.
- Cast steel, *see also* Castings, steel.  
     aging of, 203
- Castability, 58
- Castings, steel, 22, 316  
     Al in, 159  
     alloying elements in, 317  
     and wrought steel, similarity of alloy effects in, 319  
     annealing, 100  
     annealing *vs.* normalizing, 4  
     Cr, 5%, 153  
     Cr, 16-20%, 153  
     Cr, 25%, 153  
     Cr, martensitic, 153  
     Cr-Cu-Mn, 107
- Castings, Cu-Mn, 106  
     Cu-Mn-V, 106  
     Cu-Si, 212  
     Cu-Si-Cr, 108  
     Cu-Si-Mn, 212; 106  
     heat treatment, 316  
     high-C, 24  
     low-C, 23  
     Mn-Cu-Mo, 108  
     Mn-Cu-Ti, 108  
     Mn in pearlitic, 74  
     Mn-Mo, pearlitic, 76  
     Mn-Ni, pearlitic, 75  
     Mn-Ti, pearlitic, 79  
     Mn-V, pearlitic, 77  
     size effect in, 7  
     Ni, 122  
     Ni-Cr-Mo, 128  
     Ni-Mn, 122  
     Ni-Mo, 127  
     Ni-V, 127  
     normalizing, 135  
     "regular" grade, 135  
     Si, 89  
     specifications for, 23  
     V, 162  
     welding, 318
- Cementite, absorption of, by austenite on heating, 88  
     definition, 74  
     hardness of, 156  
     properties, 78  
     spheroidal, 125
- Cerium in steel, 161
- "Chapmanizing," 354  
     rate of case formation in, 355
- Charcoal, carburizing with, 266  
     gas producers, 322  
     packing, for protection of surface, 332
- Charpy impact testing, 48  
     conversion to Izod, 52
- Chisel steels, 441
- Chromium, and Mn in steel, 135  
     as carbide former, 130  
     as ferrite strengthener, 130  
     carbides, 133  
     carburizing steels, 137  
     critical cooling rate, 135  
     effect in steel, 155  
     effect on critical temperatures, 132

- Chromium, gamma loop with, **131**  
 in Hadfield Mn steel, **403**  
 in heat-resistant steel, **337**  
 in high-speed steel, **457**  
 in nitriding steel, **295**  
 increases corrosion resistance, **211**  
 tool steel, **140**  
*vs.* C content in steel, **135**
- Chromium-aluminum-molybdenum steel for nitriding, **298**
- Chromium cast steel, **153**
- Chromium-copper-manganese cast steel, **107**
- Chromium-copper-steel, **253**
- Chromium-manganese steel, **243**  
 austenitic, **373**  
 18 Cr, 8 Mn, **416**  
 spring steel, **251**
- Chromium-manganese-silicon steel, **87**, **252**  
 for low-temperature service, **383**
- Chromium-manganese-vanadium steel, **237**, **243**, **245**, **258**
- Chromium-molybdenum steel, **259**  
 carburizing steel, **282**  
 Cr 5, Mo, 0.50%, **363**  
 creep, **362**  
 for carburizing, **281**  
 tubing, **365**
- Chromium-molybdenum-nickel steel, **227**
- Chromium-molybdenum-silicon steel, **366**
- Chromium-molybdenum-vanadium steel, for large normalized sections, **266**  
 for nitriding, **296**  
 secondary hardening in, **215**
- Chromium-nickel-manganese steel, **223**
- Chromium-nickel steel, **224**, **235**, **238**  
 austenitic, **407**  
 for cold treatment, **163**
- Chromium-nickel-vanadium steel, **238**
- Chromium-silicon steel, **252**  
 valves, **375**
- Chromium steels, bearings, 17.5% Cr, **151**  
 Cr 18, Ni 8, *see* 18: 8 steel.  
 complex, **139**  
 corrosion resistance, **143**
- Chromium steels, cutlery grade, **151**  
 dies, 12-17% Cr, **151**  
 grain growth, **132**  
 high, **142**  
   nitriding, **296**  
 high-C, high-Cr tool and die steel, **439**  
   Ni, Mo, and V in, **440**  
 machinability, **152**  
 magnetic, **464**  
 P in, for creep resistance, **364**  
 sheet and strip, **160**  
 S.A.E. 51000, 52000 series, **136**  
 stainless, plain Cr, **148**  
 tubes, **144**  
 turbines, **149**
- Chromium steels 0.25% Cr, **137**
- Chromium steels 0.50% Cr, **138**
- Chromium steels 1-2% Cr, bearing steels, **141**
- Chromium steels 5% Cr, embrittlement, **365**
- Chromium steels 5% Cr, 0.50% Mo, **146**
- Chromium steels 12-14% Cr, hardening, **142**
- Chromium-titanium steel, **369**
- Chromium-tungsten steel, **363**  
 oil-hardening tool steel, **441**  
 V in, **441**  
 hot-work steels, **441**
- Chromium-vanadium steel, **254**  
 for carburizing, **281**  
 springs, **304**
- Chromizing, **372**
- Cladding, **371**
- "Cloudburst" hardness testing, **10**
- Coarsening temperatures in tool steel, **435**
- Cobalt, in high-speed steel, **457**  
 in precipitation hardenable alloys, **216**  
 in W magnet steels, **466**
- Cold treatment, **161**
- Cold working, free-cutting metals, **57**  
 tempering steel after, **199**
- Columbium, in steel, **160**  
 in 18: 8 steel, **411**
- Composition hardness, **177**
- Compressive strength, **31**

- Conduction in heating, 414
- Constructional steels, **3**
  - factors in selection, **321**
- Continuous furnaces, 409, 424, 429
- Contraction in tensile tests, 25
- Controlled atmospheres, 292
  - air-gas ratio in preparing, 325
  - CO-CO<sub>2</sub> ratio in, 307
  - carburizing and decarburizing tendencies, 328
  - effect of increased methane in, 325
  - effect on C and alloy steels, 304
  - flue gas as source of, 329
  - for heat treating, 292
  - for heating high-speed steel, 294
  - free oxygen in, 329
  - furnaces for, 302
  - gases for, 303
    - cost of, 322
    - drying, 328, 335
  - moisture effect in, 311, 317
  - oxidizing and reducing atmospheres, 292
- Convection, forced, in furnaces, 415
  - in heating, 413
- Cooking, analogy to heat treatment, 421, 432
- Cooling, curves, in various quenching media, Fig. 15, 171
  - rates, 152
    - after annealing, 102, 115
    - after drawing, 196
    - air, 131
    - for martensite and pearlite, 156
    - importance of control of, 427
  - slow, effect on structure, 90
  - of rails, **37**
  - time-temperature, relations in, 422, 425
- Copper, cast steel for crankshafts, 213
  - effect in steel, **110**
  - effect on critical temperatures, **111**
  - graphitizing effect, **100**
  - in carburizing steel, **106, 281**
  - in 18:8 steel, **413**
  - in free-cutting Cr steels, **54**
  - in Hadfield Mn steel, **405**
  - in quenched and tempered steel, **102**
  - in steel castings, **106**
- Copper, increases corrosion resistance, **210**
  - Ni replaced by, **227**
  - solubility of, in ferrite, 210
  - strengthening effect, **212**
  - substitution of, for Ni, **101**
- Copper cast steel, **106**
- Copper-chromium steel, **253**
- Copper-chromium-manganese cast steel, **107**
- Copper-manganese-molybdenum cast steel, **108**
- Copper-manganese steel for large sections, **246**
- Copper-manganese-titanium cast steel, **108**
- Copper-manganese-vanadium cast steel, **106**
- Copper-silicon-chromium cast steel, **108**
- Copper-silicon-manganese cast steel, **106**
- Copper-silicon-manganese steel castings, 212
- Copper-silicon steel castings, 212
- Copper steels, **99**
  - precipitation hardening, 209; **100**
  - surface checking of, 213; **100**
  - avoidance of, by Ni, 213
- Core properties of carburized steel, *see* Carburized steel.
- Cored grains, 84
- Corrosion fatigue, 43
- Corrosion resistance, at high temperatures, **337**
  - Cr steel, **143**
  - effect of Cu and P, **99**
  - high-yield strength, low-alloy steel, **211**
  - low-alloy steel, **209**
  - Mo steel, **171**
- Corrosion testing, 43
- Cost of alloying elements, **321**
- Cracking in quenching, 178
- Crackless plasticity, 41
- Creep, definition, **344**
  - grain size, effect of, on, **355**
  - killed *vs.* rimmed steel, **356**
  - of C steels, discrepancies in, **337**
  - resistance, 141

- Creep, resistant steel, **364**  
  strain hardening in, **346**  
  temperature effect of, on rate, **349**  
  testing, **32**  
  vs. grain-coarsening propensities, **245**
- Critical cooling rate, **156, 164**  
  Cr steel, **135**  
  definition, **167**  
  effect of alloying elements on, **184**  
  effect of Mn, Ni, Si, Cr in 0.40% C steel, **191**  
  in Mn steel, **63**  
  large sections, **221**  
  vs. grain size, **238**  
  vs. quenching media, Fig. 16, **172**
- Critical points, **81**  
  determination of, **88**  
  merging of, **86**
- Critical strain, grain growth after, **122**
- Critical temperatures, **7**
- "Crystallization," misnomer for fatigue, **33**
- Cupping tests, **56**
- Cutlery steel, **151**
- Cyanide case, depth of, effect of time and temperature on, **346**  
  nitrogen in, **345**
- Cyanide hardening, mechanism of, **346**  
  of steel surfaces, **345**
- Damage line endurance testing, **41**
- Damping, **56**
- Decarburization, **306, 312**  
  by hydrogen, **311**  
  by moisture, **297, 311**  
  effect of, on springs, **296**  
  effect on endurance limit of springs, **310**  
  high-C steel, **332**  
  high-speed steel, **450, 452**  
  in annealing, **120**  
  in carburizing process, **264**  
  tendencies in controlled atmospheres, **328**  
  vs. scaling, **293**
- Decarburized surface, effect of, on nitriding, **329, 353**
- Deep-hardening steels, effect of grain size, **237**
- Deformation, permissible, at high temperatures, **346**
- Depth hardening, effect of alloying elements on, **184**  
  in Mn steel, **63**  
  Ni steel, **115**  
  of Mo and W, **169**
- Die steels, **420**  
  intermediate alloy, **433**
- Dies, graphitic steel for, **94**
- Diffusion, **150**  
  in annealing, **107**  
  in carburizing, **257, 260, 268**  
  in steel, **106**  
  of C and N in cyaniding, **349**
- Dimensional changes in quenching, **177**
- Directional properties, **93**
- Distortion, in carburized steel, **291, 333**  
  in heat-treating carburized steel, **252, 289**
- Double annealing, **112**
- Down-step creep testing, **376**
- Draw temperatures, effect of, **195, 197**  
  range, **193**
- Drawing, effect of time, **195**  
  to soften, **192**
- Dry gas producers, **322**
- Drying gases for controlled atmospheres, **328, 335**
- Ductility, **25**  
  directional, **30**  
  effect of toughening temperature on, **194**  
  engineering requirements, **28**  
  of alloy steels, **326**  
  transverse, **31**  
  treatment for strength, **114**
- Dynamic strength, **80**
- Dynamic tests, **8, 34**
- E**
- 18 : 8 steel, additions for machinability, **412**  
  alloying elements in, **412**  
  Co in, **416**

- 18 : 8 steel, cold working, **409**  
   embrittlement, **411**  
   properties, **409**  
   stabilization, **407, 411**  
   substitution of Ni by Mn in, **413**  
   tonnage, **417**  
   welding, **410**  
 Elastic limit, **24**  
 Elastic modulus, **27**  
 Electric eye in electric heating for forging, **405**  
 Electric furnaces, **394**  
 Electric heating, **182**  
   effect on scaling, **298**  
   for forging, **386**  
   for surface hardening, **366**  
   steel for, **368**  
 Electric sheet (Si steel), **86**  
 Elements, symbols for, **3**  
 Elongation, in bend tests, **55**  
   in tensile test, **25**  
 Embrittlement, austenitic Cr-Mn steels, **413**  
   austenitic steels in welding, **398**  
   by cold work, **93**  
   by P in steel, **49, 214**  
   during high-temperature service, **338**  
   18 : 8 steel, **411**  
   hole tensile test for, **382**  
 Endurance, **33**  
   axial vs. rotary beam tests, **35, 43**  
   carburized articles, **290**  
   corrosion in, **43**  
   damage line, testing, **41**  
   limit, austempered stock, **145**  
   of commercial springs, **309**  
   spring steels, **307**  
   effect of decarburization on, **310**  
   ratio, P-alloy steel, **212**  
   testing, **35**  
   damage line, **41**  
   vs. Brinell, **36**  
   vs. tensile strength, **37**  
 Energizers for carburizing, **264, 268**  
 Equilibrium diagram, Fe-C, **81**  
 Erichsen cupping test, **56**  
 Etching, deep, **7**  
 Eutectoid composition, effect of alloying elements on, **194**  
 Eutectoid steel, **75**  
 Fast-finishing tool steels, **440**  
 Fatigue, *see* **Endurance**  
 Ferrite, absorption of, by austenite on heating, **85**  
   "beta" modification, **86**  
   definition of, **74**  
   hardness of, **156**  
   properties, **78**  
   rejection of, on cooling, **102**  
   solubility of C in, **204**  
 Ferrite strengtheners, in C tool steel, **421**  
   Cr, **130**  
   Cu, **100**  
   effects of elements, **179, 181**  
   Mn, **81**  
   Mo, **168**  
   Ni, **113**  
   P, **49**  
   Si, **85**  
 Ferrosilicon in carburizers, **268**  
 File hardness, **14**  
 Flame, luminous, in forging, **378**  
 Flame hardening of surfaces, **364**  
   steel for, **365**  
 Flue gas, as source of controlled atmospheres, **329**  
   effect on scaling, **292**  
 Ford crankshaft alloy, **213; 108**  
 Forging, burning of steel in, **386**  
   electric heating in, **386**  
   furnaces, **378**  
   slot heating vs. slugs, **384**  
   heating for, **375**  
   overheating in, **376**  
   supervision of, **407**  
 Forgings, large, normalizing, **138**  
 Fracture tests on hardenability, **232**  
 Free-cutting steels, C content of, **58**  
   cold drawing, **57**  
   for carburizing, **59**  
   for quenching, **59**  
   Pb, effect in, **54**  
   machinability, **58**  
   Mn in, **59, 69**  
   N in, **59**  
   Ni in, **113**  
   S, effect on properties of, **69**  
   S.A.E. 1100, X1100 series, **53**

- Frequency curves in statistical methods, 65
- Fuels, choice of, 408  
for heating, 391  
fluid, 391
- Furnace cooling in annealing, 104
- Furnaces, *see also* individual types, e.g.,  
Forging furnaces.  
choice of, 409  
circulation of atmospheres in, 401  
for controlled atmospheres, 302  
designs, 381  
efficient *vs.* effective, 392  
insulation, 394  
loading for uniformity, of heating, 413  
storage of heat in, 394  
temperature of, *vs.* charge temperature, 413
- Fused baths for heating steel, 342
- Gamma iron (austenite), 86  
face-centered structure, 86
- Gamma loop, formation, 180  
formed by Mo and W, 168  
formed by various elements, 183  
by Cr, 131  
by P, 52  
by Si, 84  
by V, 162
- Gas-air ratio in preparing controlled atmospheres, 325
- Gas carburizing, 271  
furnaces for, 271
- Gases, dewpoint and humidity, 335  
in metals, 5  
mixed, carburizing with, 274
- Gears, distortion in, 289
- Generator gas, 323  
adjustment for hardening, 331  
composition, 325
- Grain boundaries, depletion during carbide separation, 397
- Grain coarsening, 166  
by overheating, 97  
in normalizing, 328
- Grain growth, after critical strain, 122  
in carburized steel, 279
- Grain growth, in Cr steel, 132  
in Ni steel, 121  
in V steel, 163  
inhibition of, by solid particles, 236  
restrainers, V, Al, Ti, Zr, action of, 180
- Grain size, 219  
A.S.T.M. standards, 226  
beyond  $A_{c1}$ , 90  
coarsening on heating, 90  
control, in low alloy steel, 212  
in Mn steel, 78  
methods for, 234  
V and Al additions for, 220, 234  
effect in tool steel, 296  
effect in low-temperature, service, 382  
effect of, on deep and shallow hardening steels, 237  
on type of heat treatment, 281  
effect of hot-work on, 91  
effect of V on, 186  
effect on creep, 355  
effect on impact, 52  
for deep drawing steels, 248  
importance of, 246; 333  
in carburizing steels, 253, 281  
in tool steel, 435  
"inherent," 166  
misnomer, 237  
other standards, 228  
value of, 242  
*vs.* Al additions, 220  
*vs.* critical cooling rates, 238  
*vs.* hardenability, 219, 237; 187  
*vs.* high-temperature properties, 245  
*vs.* impact, 244  
*vs.* low-temperature properties, 247  
*vs.* machinability, 240  
*vs.* mechanical properties, 239, 243
- Grain structure, 90  
refinement of, 90
- Grains, distortion in cold working or machining, 93
- Graphite blocks for protection of surface, 332
- Graphitic steel, hardenability, 93  
machinability, 90



Graphitic steel, Mo in, **94**

Ni and Mo in, **96**

Graphitization, in C steels, **90**  
in Si-C steel, **92**

## H

Hadfield Mn steels, **399**

Hardenability, **8**

calculation from composition, **178**

effect of alloying elements on, **185**

effect of Mo on, **169**

factors affecting, **169**

fracture tests for, **232**

Mo steel, **188**

of tool steel, **243**

1 $\frac{3}{4}$ -in. rounds with 1-2% various  
elements, **189**

Sheppard penetration-fracture test,  
**232**

simple and complex steels, **177**

tests, quenching, on carburized steel,  
**232**

variation in similar steels, **233**

vs. grain size, **166, 219, 237**

Hardenability tests, hardness penetra-  
tion, **232**

surface are center method, **234**

Hardening, age, **203**

clean, **301**

depth in C steel, **167**

heating for, **147**

precipitation, *see* Precipitation hard-  
ening.

principles of, **147**

relation to annealing, **150**

secondary, *see* Secondary hardening.

surface, *see* Surface hardening.

on welding, **142**

Hardness, "cloudburst" testing, **10**

definition of, **8**

hot, **11**

Rockwell test, **12**

scales, conversion between, **15**

Shore scleroscope test, **13**

shot-blasting to evaluate, **10**

vs. tensile strength, **20**

Hardness conversions, **15**

Brinell-Rockwell B, Fig. 6, **20**

Brinell-Rockwell C, Fig. 2, **16**

Hardness conversions, Brinell-Vickers,  
Fig. 3, **17**

Rockwell C-Vickers, Fig. 4, **18**

Rockwell C-Scleroscope, Fig. 5, **19**

Hardness testing, **8**

Brinell, **8**

file, **14**

Herbert pendulum, **15**

hot, **11**

mutual indentation, **10**

precautions in, **13**

Rockwell cone test, **12**

scratch, **14**

Shore rebound, **13**

Vickers, **11**

Heat, absorption by charge, **423**

application of, **412**

flow, in steel, limited rate of, **412**  
limitations, **222**

source, **396**

storage in furnaces, **394**

Heat-resistant alloys, **370**

effect of controlled atmospheres on,  
**334**

Heat transfer, **396**

Heat treatment, **4**

carburized steel, **280**

distortion in, **283**

control of, **375**

definition, **4, 81**

for machinability, **58**

S.A.E. carburizing steels, **286**

S.A.E. steels, **202**

steel castings, **24**

Heating, for forging and rolling, **375**

for rolling, **387**

fuels for, **391**

internal, **392**

loading of furnaces for uniform, **413**

rate of, **101**

refining structures by, **148**

"saturation" in, or annealing, **95**

steels of low thermal conductivity,  
**96**

structural changes in, **87**

time of, during annealing, **102**

time-temperature relations in, **422,**  
**425**

uniformity of, **379, 413, 414**

Herbert pendulum hardness testing, **15**

- High-speed steels, **444**  
 annealing, **448**  
 atmospheres for heating, **294**  
 austenite, retained in, **446**  
 carbides in, **444**  
 composition vs. heat treatment, **451**  
 cyaniding and nitriding, **352**  
 decarburization, **295**; **450**  
 diagram for, **444**  
 14-4-1 and 18-4-1, **452**  
 grain growth in, **447**  
 hot hardness, **454**  
 Mo in, **452**  
 pack hardening, **450**  
 preheating, **450**  
 protecting by packing during heating, **295**  
 tempering, **451**  
 W in, replaced by Cr, Mo, and V, **445**  
 working, **448**  
 High-temperature properties, Mo and W steel, **170**  
 pearlitic Mn steel, **74**  
 High-temperature service, heat treatment of steels for, **337**  
 treatment for, **141**  
 High temperatures, steel for use at, **336**  
 testing, **32**  
 High-yield-strength steel, evaluation, **218**  
 not heat-treated, **207**  
 Hole-tensile embrittlement test, **382**  
 Homogenization, **131**, **135**, **150**, **164**  
 by heating, **96**, **106**  
 Hot die steels, **439**  
 Hot hardness, high-speed steel, **454**  
 testing, **11**  
 Hot work, tool steel for, **441**  
 Human element, **429**  
 in control of cooling phase, **428**  
 in heating for forging, **388**  
 Hydrocarbons as protective atmosphere, **332**  
 Hydrogen, effect on shatter cracks in rails, **37**  
 steel to withstand, at high temperatures and pressures, **374**  
 Hydrogen-methane ratio in controlled atmospheres, **306**  
 Hydrogen-water vapor ratio in controlled atmospheres, **311**  
 Hyper-eutectoid steel, definition, **76**  
 Hypo-eutectoid steel, definition, **75**  
 spheroidizing under  $Ac_1$ , **115**  
 -  
 Impact, carburized steel, **285**  
 on carburized articles, **289**  
 properties, low-temperature, **381**  
 spring steels, **307**  
 various alloy steels, **328**, **329**  
 resistance, effect of temperature on, **52**  
 of austempered stock, **144**  
 specimens, effect of dimensions of, **50**, **52**  
 test bar form, **48**  
 vs. grain size, **244**, **247**  
 Impact testing, at low temperature, **50**  
 Charpy, **48**  
 conversion of Charpy to Izod, **52**  
 effect of grain size on, **52**  
 Izod, **48**  
 notched bar, **44**  
 rail steel, **53**  
 repeated, **47**  
 service evaluation by, **53**  
 tensile, **54**  
 torsion, **54**  
 Inclusions, **6**  
 effect in high-C steel, **33**  
 Induction hardening for balancing stresses, **180**  
 Induction heating in forging, **406**  
 Insulation of furnaces, **394**  
 Internal heating, **392**  
 Iron, allotropic forms, **86**  
 "Izett" steel, **208**; **157**  
 Izod, conversion to Charpy, **52**  
 impact tests, **48**  
 K  
 Kip, **22**  
 L  
 Large sections, C steel, **30**  
 Ni and Mn in, compared, **225**  
 normalizing, **237**  
 wrought steel, **221**

- Lead, alloy baths for tempering, 192  
 baths for heating tool steel, 296, 342  
 bearing steels, 54  
 effect in free-cutting steel, 54  
 effect on mechanical properties, 56  
 in steel, 54  
 quenching into, 143, 160
- Leather as carburizer, 267
- Liquid carburizers, 348
- Low alloy steel, corrosion resistance, 209, 211  
 for riveting, 208, 217  
 for welding, 207, 209, 216  
 grain-size control, 212  
 high yield strength, Mn in, 81  
 limitation of C in welding, 213  
 yield strength, 208
- Low-temperature impact tests, 50
- Low-temperature service, grain-size effect in, 382  
 Ni steel for, 382  
 steel for, 381
- Luminous flame in forging, 378
- M**
- Machinability, annealing C tool steel for, 124  
 C tool steels, 426  
 Cr steel, 152  
 constructional steels, 332  
 Pb steels, 56  
 of carburizing steels, 254  
 Mo steel, 228  
 nitriding steels, 297  
 S.A.E. steels, 58  
 testing, 7, 57  
 vs. grain size, 240
- Magnet alloys, precipitation hardening in, 467
- Magnet steels, 460  
 C, Cr, W, 463  
 composition of, 462  
 heat treatment, 462
- Magnetic properties, effects of Si, 97
- Magnetic testing, 55  
 magnafux method for flaws, 55
- Magnetization, 460
- Magnets, permanence of and strain in, 461
- Malleable iron, 90
- Manganese, alloy or deoxidizer, 65  
 effect in steel, 81  
 effect in Si steel, 84  
 effect of varying content in heat-treated steel, 67  
 effect on hardenability, 70  
 free cutting steels, 68  
 in C tool steels, 421  
 in carburizing steel, 279  
 in Cr steel, 136  
 in 18: 8 steel, 413  
 in V steel, 163
- Manganese-carbon steel springs, 32
- Manganese cast steels, variations of, 81
- Manganese-chromium-copper cast steel, 107
- Manganese-chromium-silicon steel, 252  
 for low-temperature service, 383
- Manganese-chromium steel, 243  
 18 Cr, 8 Mn, 416  
 spring steel, 251
- Manganese-chromium-vanadium steel, 237, 243, 245, 258
- Manganese-copper steel, 246
- Manganese-copper-silicon cast steel, 106
- Manganese-copper-vanadium cast steel, 106
- Manganese-molybdenum steel, 246
- Manganese-nickel-chromium-molybdenum steel, 236
- Manganese-nickel-chromium steel, 223
- Manganese-nickel-molybdenum cast steel, 127
- Manganese-silicon-chromium steel, 87
- Manganese-silicon-nickel-chromium-molybdenum steel, 236
- Manganese steels, 210  
 Al, Ti, and V in, 73  
 austenite retention in, 73  
 austenitic, 399  
 annealing and quenching, 402  
 Cr in, 403  
 Cu in, 405  
 heat treatment, 400  
 high-Mn, structure, 401  
 Ni in, 399  
 wear resistance, 399, 406  
 work hardening, 407

- Manganese steels, C and Mn balance in, **66**  
constructional, **61**  
critical cooling rate, **63**  
depth hardening properties, **63**  
for low-temperature service, **383**  
grain growth in, **238**  
hardenableity of, **63**  
high-temperature properties, **74**  
medium, normalizing of, **238**  
non-deforming, **73**  
normalizing, **70**  
rifle barrels of, **72**  
"sensitivity," **70**  
S.A.E. X1300, X1000, T1300 series, **66**  
S, effect on properties, **69**  
temper brittleness, **70**  
2% Mn for shear blades, **71**  
welding, **68**  
yield ratio, **62**
- Manganese steels 1.5-2% Mn, **69, 81**  
Manganese steels 1.10-1.30% Mn, **71**  
Manganese steels 1.60-1.90% Mn, **72**  
Manganese-vanadium steel, **237, 243**  
welding, **214**
- Martensite, alpha, **155**  
beta, **155**  
cracks in, **143**  
definition, **155**  
hardness of, **156**  
primary, formation, **156**  
tetragonal, in magnet steels, **462**  
transformation at constant temperature, **144**
- Mass effect, *See* Depth hardening.
- McQuaid-Ehn test, **220, 222**  
limitations of, **222**
- Merit indices, **58**
- Methane-hydrogen ratio in controlled atmospheres, **306**
- Microscopic examination, **6**
- Modulus of elasticity, **27**
- Moisture, effect of, in controlled atmospheres, **311, 317**  
produced in controlled atmospheres, effect of, **319**  
removal of, **335**
- Molybdenum, carbide, stability, **170**  
depth hardening effect, **169**  
effect on grain growth, **168**
- Molybdenum, effect on temper brittleness, **215; 228**  
for depth hardening, **188**  
gamma loop formation, **168**  
in carburizing steel, **279**  
in Cr magnet steel, **464**  
in graphitic Si-C steel, **95**  
in Mn pearlitic steel castings, **76**  
in Mn steel, **70**  
in nitriding steels, **295**  
in pearlitic Cr steel, **144**  
in precipitation hardenable alloys, **215**
- Molybdenum-aluminum steel, springs, **173, 302**
- Molybdenum-carbon steel, creep, **362**
- Molybdenum-chromium-aluminum steel, for nitriding, properties, **298**
- Molybdenum-chromium-nickel steel, **227**
- Molybdenum-chromium-silicon steel, **366**
- Molybdenum-chromium steel, **259**  
5% Cr, 0.50% Mo, **363**  
high Mo content, **261**  
normal Mo, **263**  
tubing, **365**
- Molybdenum-chromium-vanadium steel, for large normalized sections, **266**  
for nitriding, **296**
- Molybdenum high-speed steels, **452**  
decarburization, **452**  
structure, **453**
- Molybdenum-manganese steels, **246**
- Molybdenum-nickel-manganese cast steels, **127**
- Molybdenum-nickel-silicon steel, **243**
- Molybdenum-nickel steel, **241**
- Molybdenum-silicon-manganese-copper tool steel, **442**
- Molybdenum steels, 0.20% Mo, 0.65% C, **173**  
0.36% Mo, **172**  
0.50% Mo, **175, 338**  
creep, **356**  
1.38% Mo, **175**  
carbide forming effects, **168**  
carbides in, **363**

- Molybdenum steels, complex, **175**  
 corrosion resistance, **171**  
 creep properties, **339**  
 decarburization, **171**  
 ferrite strengthening effects, **168**  
 machinability, **228**  
 quenched and tempered, **172**  
 vs. C steel at high temperatures, **361**  
 Molybdenum-vanadium steels in guns, **267**  
 Muffle furnaces, **404**  
 Mushet steel, **440**  
 Mutual indentation hardness testing, **11**
- N**
- Natural gas for carburizing, **274**  
 Network, cementite, annealing to absorb, **130**  
 of ferrite or cementite, **90**  
 Nickel, cast steel, **122**  
 effect in normalized steel, **113**  
 effect in quenched and tempered steel, **115**  
 effect in steel, **128**  
 effect on critical temperatures, **120**  
 effect on eutectoid ratio, **121**  
 -free steel, **227**  
 in carburizing steel, **122, 279**  
 in 18: 8, **408**  
 in graphitic Si-C steel, **96**  
 in Hadfield steel, **403**  
 in nitriding steel, **296**  
 in Si steel, **89**  
 increases corrosion resistance, **211**  
 replacement by Cu, **227**  
 replacement by Mn, **225**  
 substitution of, by Cu, **101**  
 Nickel-chromium-manganese-molybdenum steel, **236**  
 Nickel-chromium-manganese steel, **223**  
 Nickel-chromium-molybdenum cast steel, **227, 228, 238, 436**  
 Nickel-chromium-molybdenum-vanadium steel, **230, 236, 436**  
 Nickel-chromium steel (*see also* 18: 8),  
 austenitic, **224, 407**  
 battering tools, **436**  
 cast, **235, 238**  
 Nickel-chromium steel, critical points, **238**  
 for carburizing, **281**  
 property charts, **238**  
 temper brittleness, **206; 228**  
 Nickel-chromium-vanadium steel, **228, 238**  
 Nickel-manganese cast steel, **122**  
 Nickel-manganese-molybdenum cast steel, **127**  
 Nickel-molybdenum cast steel, **127**  
 Nickel-molybdenum steel, carburizing, **281**  
 Nickel-molybdenum-vanadium steel, for turbine forgings, **140**  
 Nickel-silicon-molybdenum steel, **243**  
 Nickel-silicon steel, **243**  
 Nickel steels, **113**  
 austenite retention, **121**  
 austenitic, **398**  
 critical cooling rate, **116**  
 depth hardening, **115**  
 for large forgings, **115**  
 for low-temperature service, **382, 384**  
 hardenability, **119**  
 heat-treatment temperatures for, **121**  
 3½ vs. 1% Ni, **120**  
 4.5% Ni, quenching, **183**  
 5% Ni, effect of cooling rate, **176**  
 Nitrided case, hardness and toughness, **355**  
 properties, **359**  
 Nitrides, precipitation hardening by, **207**  
 Nitriding, **350**  
 carburizing processes, **353**  
 containers for, **360**  
 effect of alloy content of steel in, **352**  
 effect of temperature on, **355**  
 effect on decarburized surface, **329**  
 growth in, **358**  
 high-C, high-Cr steel, **296**  
 preparation for, **294**  
 preparation of surface for, **352**  
 protection against, **359**  
 rate of, **355**  
 steels, **294**  
 case thickness, **294**  
 precipitation hardenable, **214**  
 time required, **357**

- Nitriding, tool steel, 352  
     warping in, 358  
 Nitrogen, as protective atmosphere, 303  
     diffusion of, in cyaniding, 349  
     in free cutting steels, 59  
 Non-deforming steel, 179; 436  
     Mn in, 73  
 Normal steel in carburizing, 219  
 Normalizing, 103, 131  
     atmospheres for, 299  
     effect of rate of cooling in, on prop-  
         erties, 105  
     effect of V, 162  
     for structure stability, 141  
     forgings, 138  
     grain coarsening in, 238  
     initial structure for, 106  
     large sections, 237  
     prior to carburizing, 254  
     release of stress in large sections, 237  
     sheet and strip, 134  
     single pieces *vs.* a mass, 427  
     steel castings, 135; 25  
     structures produced by, 105  
     temperature uniformity in, *vs.* pack  
         annealing, 417  
     tempering after, 139  
     *vs.* annealing, 141  
         cast steel, 25  
 "Normalloy," 245  
 Notch effect, 38  
 Notched bar tests, endurance, 40  
     impact, 44  
     slow bend, 45  
 Notch sensitivity in high-C steel, 33
- O
- Oil, for carburizing, 274  
     vapor, as protective atmosphere, 332  
 Olsen cupping test, 56  
 Overheated steel, refining, 109  
 Overheating in forging, 376  
 Overstressing in fatigue, 33  
 Oxides, precipitation hardening by, 207  
 Oxidizing atmospheres, 292  
 Oxygen, absorption in box carburizing,  
     266  
     free, in controlled atmospheres, 329
- Patenting, 142  
 Pearlite, C content, 132  
     definition of, 75  
     effect of annealing, cooling rate on,  
         115  
     emulsified, 133  
         as initial structure for spheroidiz-  
             ing, 126  
     hardness, 156  
     lamellar (*see also* Troostite, pri-  
         mary), 132  
     variation of impact with fineness,  
         141  
     variation of strength with fineness,  
         133  
     properties, 78, 133  
 Pendulum hardness, 15  
 Permanent magnet steels, 460  
 Phosphorus, avoidance in carburizing  
     steel, 281  
     effect in steel, 49  
     effect on normalizing temperature, 50  
     embrittlement by, in steel, 49  
     for increased yield strength and cor-  
         rosion resistance, 51  
     high-temperature effects, 364  
     in Cr steel, creep-resistant, 364  
     in Mn steel, for low-temperature use,  
         385  
     in steel, endurance ratio, 212  
     increases corrosion resistance, 211  
     limitation, 214  
     segregation of, 50  
 Plate, carburizing, 275  
 Precipitation hardenable alloys, of Fe,  
     Mo, W, Co, V, 215  
 Precipitation hardening, 203  
     agglomeration of particles and soften-  
         ing after, 205  
     alloys, 215  
         for permanent magnets, 467  
         solubility in, 205  
     at high temperatures, 337  
     austenitic steels by carbide separa-  
         tion, 397  
     by Ni-Al compounds, 213  
     in nitriding steels, 214  
     magnetic alloys, Co, W, Mo, etc., in,  
         467

Precipitation hardening, of Cu steel, 209; **100**  
 of V steel, **165**  
 Pressure quenching, **13**  
 Price of alloying elements, **321**  
 Probability methods in testing, **65**  
 Proof load, **24**  
 Propane for gas carburizing, **273**  
 Proportional limit, **24**; **340**  
 Pyrometer readings *vs.* charge temperature, **413**, **419**

## Q

Quench, regenerative, **281**  
 single and double, for carburized steel, **282**  
 -toughening, **177**, **199**  
 Quenching, brine, **173**  
 C steel, **43**  
 in oil *vs.* water, **26**  
 caustic soda, **173**  
 circulation in, **170**, **175**  
 cracking and warping in, **178**  
 dimensional changes in, **177**  
 effect of surface condition in, **174**  
 hardening, "T max." in, **165**  
 in spray or pressure, **176**  
 mass effect in, **165**  
 media, **167**  
 severity of, **182**  
 temperature of, **175**  
*vs.* toughening temperature, **197**  
 mild, **182**  
 oil, **171**  
 requirements for, **173**  
 pressure, **176**  
 single pieces *vs.* a mass, **428**  
 size effect in, **165**  
 steel castings, **333**  
 stresses, **177**, **180**  
 constructional steels, **333**  
 sulphuric acid, **173**  
 surface hardening by, **364**  
 temperature, effect on critical cooling rate, **165**, **177**  
 temperatures below that of heating, **182**  
 time, **177**  
 rail ends, **39**  
 Quenching, vapor jacket in, **170**  
 water, **170**

## R

Radiant-tube furnaces, **401**  
 Radiation, **397**  
 in heating, **414**, **418**  
 Radiography, **55**  
 Rail ends, heat treatment, **199**; **38**  
 Rail steels, **35**  
 impact, **53**  
 medium Mn, **68**  
 shatter cracks in, **36**  
 transverse fissures in, **36**  
 wear, **37**  
 Rails, austenitic Mn steel, **400**  
 heat treatment by quenching, **37**  
 slow cooling of, **37**  
 Raw materials in steel making, cost of, **34**  
 Razor steels, **433**  
 Red hardness, high-speed steel, **454**  
 Reducing atmospheres, **292**  
 Reduction of area in tensile tests, **25**  
 Refining, overheated steel, **109**, **110**  
 structure by heating, **148**  
 Refractories, insulating, **395**  
 Relaxation testing, **377**  
 Resistor alloys, effect of controlled atmospheres on, **400**  
 Resistors, effect of controlled atmospheres, **400**  
 for electric furnaces, **399**  
 non-metallic, **404**  
 Rimmed steel, **356**  
 Rockwell cone test, **12**  
 Rockwell C *vs.* tensile, Fig. 7, **21**  
 Rolling, heating for, **387**  
 Rustless iron, *see* Stainless steels.

S curves, for pearlite and martensite transformations, **160**  
 S-N curve, endurance testing, **36**  
 Salt baths, activated, for cyaniding, **349**  
 cyanide in, **344**  
 for heating steel, **342**  
 for tempering, **192**  
 materials for, **343**  
 "Saturation" in heating, **95**

- Sealing, 378  
 by  $\text{CO}_2$ , 310  
 effect of flue gas on, 292  
 electric heating, effect on, 298  
 in annealing, 120  
 prevention of, 292  
 "smoke" burners for prevention of, 292  
 S in gas, effect on, 335  
 time-temperature effects on, 297  
 vs. decarburization, 293  
 Scleroscope hardness test, 13  
 Scratch tests, 14  
 Screw stock, cold-drawn, tempering, 200  
 Secondary hardening, cases due to precipitation hardening, 215  
   in high-speed steel, 446  
   of V steels, 215; 166  
 Segregation, 6  
   in P steel, 50  
 Selenium, in free-cutting steel, 53  
   in nitriding steels, 297  
 Shallow hardening steels, effect of grain size, 237  
 Shatter cracks, in rail steel, 36  
 Shearing strength, 31  
 Sheet steel, annealing and normalizing, 134  
   Cr, 150  
 Shock testing, 44  
 Shore hardness, 13  
 Sigma phase, in Cr-Ni-Mn steel, 414  
 Silicon, as ferrite strengthener, 97  
   effect in steel, 97  
   effect on critical temperatures, 84  
   effect on hardenability, 86  
   in carburizing steel, 284  
   in Cr-Mo steel, 367  
   in 18:8 steel, 413  
   in graphitic steel, 90  
   in steel castings, 89  
   in tool steel, 89  
 Silicon cast steel, 8  
 Silicon-chromium-manganese steel for low temperature service, 383  
 Silicon-chromium-molybdenum steel, 366  
   for tough tools, 89  
 Silicon-chromium steel, 252  
   valves, 375  
 Silicon-copper-manganese cast steel, 106  
 Siliconizing, 372  
 Silicon-manganese-chromium steel, 252  
 Silicon-manganese steel, 84  
   springs, 304  
 Silicon-molybdenum-nickel steel, 243  
 Silicon-nickel steel, 89, 243  
 Silicon spring steel, 87  
 Silicon steels, 84, 210  
   carburization and decarburization, 85  
   grain growth, 86  
   high-temperature properties, 86  
   quenched, 87  
 Size effect, in quenching, 165  
   in steel castings, 6  
   on properties, 5  
 Slow bend tests, notched bar, 45  
 "Smoke" burners for prevention of scaling, 293  
 S.A.E. steels, carburizing steels, 285; 274  
   creep data, 380  
   for low temperatures, 384  
   free-cutting steels, 53  
   heat treatment suggested for, 286; 202  
   list of C steels, 5, 197  
   Mn steels, 66  
 S.A.E. X1000, 66  
 S.A.E. X1015, 332  
 S.A.E. 1020, 15  
   carburized steel, case and core properties, 286-289  
   properties, 43-44  
 S.A.E. 1030, 19  
 S.A.E. 1035, properties, 45  
   water quenched, 41  
 S.A.E. 1040, 27, 332  
 S.A.E., 1045, properties, 46  
   water quenched, 41  
 S.A.E. 1050, 30  
 S.A.E. 1112 free cutting steel, 53  
 S.A.E. T1300, 66  
 S.A.E. X1300, 59, 66  
 S.A.E. X1314, 332  
 S.A.E. X1314-X1340, 53  
 S.A.E. 1330, 328, 329  
 S.A.E. T1330, 332



- S.A.E. T1340, **332**  
 S.A.E. 2000 series, **114**  
 S.A.E. 2015, **122**  
 S.A.E. 2115, **122**  
   carburizing steel, core properties, **277-278**  
 S.A.E. 2315, **122**  
   carburized steel, case and core properties, **286-289**  
 S.A.E. 2330, **328-329**  
 S.A.E. 2330 and 2340 compared, **119**  
 S.A.E. 2340 and 1045 compared, **115**  
 S.A.E. 2345, **328, 329**  
 S.A.E. 2512, carburizing steel, core properties, **277-278**  
 S.A.E. 2515, **277-278**  
 S.A.E. 2512, **286**  
 S.A.E. 3115, **332**  
   carburized steel, case and core properties, **286-289**  
   carburizing steel, core properties, **277-278**  
 S.A.E. 3120, **271**  
 S.A.E. 3130, **62; 239, 328, 329**  
 S.A.E. 3140, **332**  
 S.A.E. 3145, **328, 329**  
 S.A.E. 3240, **233, 239, 328, 329**  
 S.A.E. 3312, **277-278**  
 S.A.E. 3415, **277-278**  
 S.A.E. 3435, **233**  
 S.A.E. 4100, **254**  
 S.A.E. 4120, **277, 278, 291**  
 S.A.E. 4130, **328, 329, 332, 375**  
   normalized tubing, **263**  
 S.A.E. X4130, **263**  
 S.A.E. 4140, **332**  
 S.A.E. 4330, **358**  
 S.A.E. 4340, **328, 329**  
 S.A.E. 4615, carburized hardness of case, **271**  
   carburized steel, case and core properties, **286-289**  
   carburizing steel, core properties, **277-278**  
   for carburizing, **291**  
 S.A.E. 4620, **332**  
 S.A.E. 4645, **328, 329**  
 S.A.E. 4820, **286-289**  
   carburizing steel, core properties, **277, 278**  
   S.A.E. 5100, **138**  
   S.A.E. 5130, **332**  
   S.A.E. 5135, **332**  
   S.A.E. 6100, **254**  
   S.A.E. 6115, **286-289**  
     carburizing steel, core properties, **277-278**  
   S.A.E. 6120, **332, 328, 375**  
   S.A.E. 6130, **59; 328, 329**  
   S.A.E. 6145, **328, 329**  
   S.A.E. 7260, **375**  
   S.A.E. 9255, **87**  
   S.A.E. 51000, **136**  
   S.A.E. 52000, **136**  
 Sodium cyanide baths for surface hardening, **348**  
 Solid solutions, **82**  
 Sorbite, **186, 192**  
   definition of, **133**  
   stages in production of, **187**  
   structure of, **133**  
 Spark testing, **5**  
 Spheroidization, **132, 141**  
 Spheroidizing, **124**  
   anneal, **110**  
   high-C steel, **130**  
   hypoeutectoid steel under  $A_{c1}$ , **115**  
 "Split" transformation, **157**  
 Stability of structure, normalizing for, **141**  
 Stabilization of austenitic steels, **398**  
 Stainless iron, *see* Stainless steels.  
 Stainless steels, **148, 323**  
   bars, prices of, **323**  
   high Cr or Ni, **416**  
   springs, **304**  
   tonnage of, **417**  
 Static strength, **80**  
 Static tests, **8, 34**  
 Statistical methods, use in testing, **59**  
 Stead's brittleness, **121; 14**  
 Steel castings, *See* Castings, steel.  
 Steels, charts of properties, limitations of, **40**  
   definition of, **2, 74**  
   for high-temperature use, **336**  
   for low-temperature use, **381**  
   hydrogen resistant, at high temperatures, **374**

- Steels, killed, grain size, **357**  
rimmed, creep of, **356**  
slowly cooled structure, 76  
structure, 2  
for high-temperature use, **336**  
welding, **215**  
Stellite tools, **458**  
Strain hardening, in creep, **346**  
Strength, calculation from compositions, **178**  
in large sections, **221**  
Stress, induced, used to resist applied stress, **183**  
internal, effect on impact and endurance, **180**  
relief, annealing, of welds, **142**  
by tempering, **140, 180, 187**  
of turbine forgings, **140**  
strain-diagrams, **22, 25**  
Stress concentration, **38**  
behavior of steel under, **333**  
Stresses, balancing of, **181**  
set up in quenching, **177**  
Stress-raisers, **38**  
Stretcher strains, **208**  
in deep drawing, *vs.* grain size, **248**  
stabilization against, **208**  
Strip steel, Cr, **150**  
normalizing, **134**  
Structure affect by annealing, **96**  
Structures, characteristic of transformation temperature, **157**  
produced in heating through the critical range, **148**  
Sulphur, effect on rate of diffusion of C in steel, **281**  
in carburizing steel, **281**  
in free cutting steels, **53**  
in gas, effect on scaling, **296, 335**  
in nitriding steels, **296**  
segregation, **6**  
Surface hardening, by electric induction heating, **366**  
by quenching, **364**  
grain size in, **369**  
"Tocco" process, **366**  
Surface reactions in carburizing, **251**  
Symbols for chemical elements, **3**  
"T max." in Mo steel, **169**  
in quench hardening, **165**  
relation to grain coarsening, **238**  
Tantalum, in steel, **161**  
Temper brittleness, **197, 206**  
effect of Mo on, **206**  
in Ni-Cr steel, **206; 227**  
Temper colors, **187**  
limitations of tempering by, **188**  
"Temper" of C tool steels, **424**  
Temperature, control of, **400**  
for annealing, **89, 100**  
gradient in heating steel, **413**  
of furnace *vs.* charge, **413**  
Tempering, after normalizing, **139**  
air circulation in, **192**  
C tool steels, **429**  
carburized steel, **284; 274**  
cold worked steel, **199**  
colors for tools, **431**  
cooling after, **189**  
definition of, **186**  
effect of time, **195**  
hardening strains, **187**  
liquid baths for, **190**  
on plates or in sand, **191**  
residual stress removed by, **200**  
resistance of Mo and W steel, **170**  
resistance to, **193**  
Ni, Si, Cr, and W steel compared, **191, 194**  
Mo steel, **192**  
steel castings, **25**  
Tensile *vs.* Brinell, Fig. 8, **21**  
Tensile impact testing, **54**  
Tensile strength, *vs.* endurance limit, **57**  
*vs.* hardness, **20**  
*vs.* Rockwell C, Fig. 7, **21**  
Tensile testing, **20**  
applications of, **28**  
definition of, **20**  
dimension of test bars, **22**  
precautions in, **26**  
rate of loading, **26**  
Testing (*see also* individual tests, e.g., Tensile testing; Magnetic testing), **460**  
objects of, **4**  
use of statistical methods, **59**

- Thermal analysis, 88  
 Thermal conductivity, relation to heating rate, 96  
 "Time" quenching, 177, 199  
 Time-temperature relations, in heating, 96  
   in heating or cooling, 422, 425  
 Time-to-rupture tests, 342  
 Titanium, as carbide stabilizer, 374  
   in cast steel, 160  
   in Cu cast steel, 108  
   in 18:8 steel, 411  
   in Mn steel, 73  
   in pearlitic Mn steel castings, 79  
 Titanium-chromium steel, 369  
 "Tocco" process of surface hardening, 366  
 Tool steels, annealing, 126  
   "body" of, 243  
   "brittle timbre," 295  
   carbide formers in, 438  
   carbon, 420  
     carbides, in, 421  
     classification by grade, 422  
     heat-treatment limitations, 421  
     heating time, rate and temperature, 428  
     initial structure, 426  
     machinability, 426  
     Mn in, 421  
     Navy specifications, 424  
     properties of, 425  
     quenching temperature, 430  
     requirements for, 420  
     "temper" of, 424  
     tempering of, 429  
     toughness, 429  
     V in, 429  
     water-hardening, Mn in, 425  
 Cr in, 140  
   effect of grain size, 296  
   for razors, 433  
   hardenable, 243  
   intermediate alloy, 433  
     carbides in and toughness of, 433  
     ferrite strengtheners in, 436  
     high-C, high-Cr, 439  
     grain size and coarsening temperature, 435  
 Tool steels, intermediate alloy, Ni-Cr, and Ni-Cr-Mo, 436  
   oil quenching, 439  
   Si and Mn in, 436  
   toughness service evaluation, 434  
   nitriding, 352  
   non-deforming, 189  
   protection of, during heating, 295  
   service evaluation, 434  
   Si in, 89  
 Torsion impact testing, 54  
 Torsional strength, 31  
 Toughness, definition of, 192  
   *vs.* C content, 198  
   *vs.* annealing, 198  
 Transformation, of martensite, at constant temperature, 143, 156  
   "split" or doubled, 157  
 Transformer steel (Si steel), 86  
 Transverse fissures in rail steel, 36  
 Transverse properties, 93  
 Troostite, definition, 154  
   etching characteristics, 193  
   hardness of, 156  
   primary, 133, 153  
     formation, 156  
     pearlitic structure, 156  
   secondary, 186  
 Tube steel, 144  
 Tubing, heat treated, 263  
 Tungsten, carbide, stability, 170  
   gamma loop formation, 168  
   in Cr-magnet steel, 464  
   in Cr steel, 363  
   in high-speed steel, 452  
   in precipitation hardening alloys, 215  
 Tungsten-chromium steel, oil hardening tool steel, 441  
 Tungsten magnet steel, 464  
   heat treatment, 465  
 Tungsten-manganese-silicon-chromium steel, 440  
 Tungsten steels, carbide forming effect, 168  
 Tungsten tool steels, 440  
 Turbine forgings, heat treatment, 140  
 Turbine shafts, composition and treatment, 237

## U

- Understressing in fatigue, 34
- Up-step testing, 376
- Uranium, high-speed steel, 458
  - in steel, 161
- Valve steels, austenitic, 375
  - automotive, 375
  - Cr-Ni-Al, 215
  - Cr-Si, 375
- Vanadium, carbide, 163
  - effect in steel, 161
  - effect on grain size and hardenability, 186
  - gamma loop from, 162
  - in carburizing steel, 284
  - in creep-resistant steel, 363
  - in high-speed steel, 452
  - in Mn pearlitic steel castings, 77
  - in Mn steel, 73
  - in medium Mn steel for low-temperature use, 383
  - in Ni steel, 116
  - in Ni-Cr steel, 228
  - in nitriding steels, 295
  - in precipitation hardenable alloys, 216
- Vanadium cast steel, 162
- Vanadium-chromium-manganese steel, 237, 243, 245, 258
- Vanadium-chromium-molybdenum steel, for large normalized sections, 266
  - for nitriding, 296
- Vanadium-chromium-nickel steel, 238
- Vanadium-chromium steel, 254
- Vanadium-copper-manganese cast steel, 106
- Vanadium-manganese steel, 163, 237, 243
- Vanadium-molybdenum steel, in guns, 267
- Vanadium-nickel steel, 166
- Vanadium steels, grain growth, 163
  - precipitation hardening, 165
  - secondary hardening, 166

- Vanadium steels, tool steels, 436
  - wrought, 162
- Vickers pyramid hardness test, 11

## W

- Warping in quenching, 178
- Water gas reactions, 311
- Wear, resistance, 57
  - testing, 57
- Weldability, 56
- Welding, austenitic steels, 398
  - C limitation in, low alloy steel, 213
  - hardening in, 142
  - low-alloy, high-yield-strength steel, 207
  - Mn steel, 68
  - steel castings, 68, 318
- White iron, 90
- Wire, C steel, heat treatment, 34
  - fatigue of, 35
  - patented, 142
  - cold drawn *vs.* heat treated, 34
- Wrought alloys, Cr-Fe, 154
- Wrought carbon steel, 4
- Wrought steels, large sections of, 221
  - V in, 162
  - vs.* cast steel, similarity of alloying effect in, 319

## X

- X magnification symbol, 78

- Yield point, 24
- Yield ratio, definition, 4
  - effect of Cu, 99
  - effect of Ni, 113
  - Si steels, 85
- Yield strength, 24
  - alloying to meet, 210
  - low alloy steel, 208

